AF
IFW

Appl. No. 10/771,522
Atty. Dkt. No. 023971-0371

IN THE UNITED STATES PATENT AND TRADEMARK OFFICE
BEFORE THE BOARD OF PATENT APPEALS AND INTERFERENCES

Appellants: Junpei OGAWA et al.

Title: HIGH-STRENGTH CONNECTING ROD AND METHOD OF PRODUCING SAME

Appl. No.: 10/771,522

Filing Date: 02/05/2004

Examiner: Vinh LUONG

Art Unit: 3682

BRIEF ON APPEAL

Mail Stop Appeal Brief - Patents
P.O. Box 1450
Alexandria, VA 22313-1450

Sir:

Under the provisions of 37 C.F.R. § 41.37, this Appeal Brief is being filed together with a credit card payment form in the amount of $500.00 covering the 37 C.F.R. 41.20(b)(2) appeal fee. If this fee is deemed to be insufficient, authorization is hereby given to charge any deficiency (or credit any balance) to the undersigned deposit account 19-0741.

REAL PARTY IN INTEREST

The real party in interest is Nissan Motor Co., LTD.

RELATED APPEALS AND INTERFERENCES

There are no related appeals and interferences.

STATUS OF CLAIMS

The following constitutes a statement of the status of all the claims, wherein all claims currently rejected (claims 1, 2, 4, 19 and 21-25) are hereby appealed:

1.	Rejected
2.	Rejected
3.	Objected To
4.	Rejected
5 – 18.	Withdrawn
19.	Rejected
20.	Allowed
21 – 25.	Rejected
26 – 28.	Withdrawn

STATUS OF AMENDMENTS

1) There are no claim amendments filed subsequent to the final rejection.

2) There is only one specification amendment that was filed subsequent to the final rejection, constituting the duplication of the ***exact*** language of a portion of originally filed claim 19 (second to last recitation) into the specification, along with the preamble "In another embodiment," in an effort to alleviate any concerns regarding an unfounded written description issue raised during the January 18, 2006, interview. This amendment was refused entry on the grounds that the phrase "in another embodiment" raises new issues that would require further consideration."

a) Proffered amendment to the paragraph starting on line 6 of page 24 of the specification, that was subsequently refused entry:

The high-strength connecting rod of this invention is a connecting rod so shaped as to have a connecting beam section, a big end, a small end and a joining section as stated above. The connecting rod has a portion of the smallest cross sectional area in its connecting beam section, a portion of the lowest fatigue strength at its big or small end, and a portion of varying fatigue strength in its joining and connecting beam sections. **In another embodiment, a portion which is the lowest in fatigue strength exists in at least one of the big and small ends, and a portion which varies in fatigue strength exists in each of the first and second joining sections and in the connecting beam sections.** The connecting rod is so made that the product of its cross sectional area and fatigue strength at cross section of its joining and connecting beam sections may be equal to or greater than the product of its cross sectional area and fatigue strength in its portion of the smallest cross sectional area in its connecting beam section. The connecting rod contains 0.73% or less of C on a mass basis (i.e., % by weight) and is so made that the cross section of each of its connecting beam and joining sections may be composed of a tempered martensitic structure or a ferritic-pearlitic structure, or a mixture of these structures satisfying relational expression or Eq. (1) given above. At least the entire cross section of its portion of the smallest cross sectional area in its

connecting beam section may be of a tempered martensitic structure. Therefore, it is possible to achieve a reduction of residual stress in its fully hardened portion and its boundary of hardening, an improvement in the fatigue strength of the connecting rod and a reduction in the weight of the part.

SUMMARY OF CLAIMED SUBJECT MATTER

Claim 1: Claim 1 recites a connecting rod which may be used, for example, to connect a crank shaft to a piston in an internal combustion engine. Referring to Figs. 1 and 2, and the specification, for example, at page 9, line 19 to page 11, line 21, the connecting rod 10 includes a connecting beam section / main body 40, a big end 20 axially opposite a small end 60, a first joining section 30 located between the connecting beam section 40 and the big end 20 to connect the beam section 40 and the big end 20, and a second joining section 50 located between the connecting beam section 40 and the small end 60 to connect the beam section 40 and the small end 60. As may be seen from Figs. 1 to 6, each of the first and second joining sections (30, 50) gradually and continuously decreases in cross sectional area toward the connecting beam section (40).

Claim 1 further recites that "each of the first and second ***joining sections* . . . *has a strength distribution in which a strength*** ***<u>increases</u>*** ***with a decrease in the cross sectional area.***" (Emphasis added.) That is, as opposed to a prior art connecting rod where strength ***decreases*** with a ***decrease*** in cross sectional area, the connecting rod has a strength distribution that changes in an ***opposite*** manner. Fig. 7, in view of Fig. 1, graphically illustrates this feature of claim 1, as is detailed on page 11, lines 22 to page 12, line 16 of the specification. This feature imparts the advantage of providing a connecting rod that is easily machined in the areas where machining must take place (*e.g.,* at/near the big and small ends which rotationally connect to the crank shaft and piston respectively), due to the relatively low strength in these locations, and a rod that has relatively high strength in locations that do not need to be machined, which correspond to sections of relatively lower cross sectional area. The connecting rod is easily machined in these areas because the hardness of the material is low. This invention provides the "best of both worlds": ease of machineability in areas which must be machined, and high strength in areas that need not be machined.

Claim 2: Claim 2 recites a connecting rod as claimed in claim 1, wherein the strength distribution is based on a proportion (%) of martensite. Fig. 9 presents a graphic

depiction of the change in the proportion of martensite with respect to location on the connecting rod (as well as depicting the cross sectional area of the rod with respect to location on the connecting rod). The specification, from page 12, line 21 to page 14, line 8, for example, details the features of claim 2.

Claim 3: Claim 3, which is allowed, recites a connecting rod as claimed in Claim 2, wherein the proportion of martensite (%) changes based on a change of the cross sectional area of each of the first and second joining sections in a manner to satisfy a relationship represented by the following formula:

$$D/D_{min} \geq 1/((1-\alpha) \times Ms/100 + \alpha)$$

where D_{min} is the minimum value of the cross sectional area of each of the first and second joining sections; and α is a value obtained by dividing a buckling stress without hardening by a buckling stress with hardening. The specification, from page 12, line 31, to page 13, line 11, for example, details the features of claim 3.

Claim 4: Claim 4 recites a connecting rod as claimed in claim 2, wherein the strength distribution is formed based on a distribution in at least one of a hardening temperature and a tempering time for each of the first and second joining sections. The specification details this feature, for example, from page 12, line 21, to page 15, line 31.

Claim 19: Claim 19 recites a novel connecting rod which also may be used, for example, to connect a crank shaft to a piston in an internal combustion engine. Referring to Figs. 1 and 2, and the specification at page 9, line 19 to page 10, line 32, page 21, line 27 to page 25, line 2 and the example on page 32, line 4 to page 33, line 15, the connecting rod 10 includes a connecting beam section / main body 40, as is detailed in claim 1, with the further feature that the connecting beam section has "a smallest cross sectional area portion which is the smallest in cross sectional area throughout the connecting rod." As with claim 1, the connecting rod also has a big end 20 axially opposite a small end 60, a first joining section 30 located between the connecting beam section 40 and the big end 20 to connect the beam

section 40 and the big end 20, and a second joining section 50 located between the connecting beam section 40 and the small end 60 to connect the beam section 40 and the small end 60. As may be seen from Figs. 1 to 6, each of the first and second joining sections (30, 50) gradually and continuously decreases in cross sectional area toward the connecting beam section (40).

Claim 19 further recites the feature of "a lowest fatigue strength portion which is the lowest in fatigue strength [that] exists in at least one of the big and small ends," along with the feature of "a variable fatigue strength portion which varies in fatigue strength that exists in each of the first and second joining sections and in the connecting beam section." Further, according to claim 19, (a) a product of (i) the cross sectional area and (ii) the fatigue strength at a cross section of ***each of the joining and connecting beam section*** is equal to or greater than (b) a product of (iii) the cross sectional area and (iv) the fatigue strength ***in the smallest cross sectional area portion in the connecting beam section***.

As with claim 1, claim 19 provides a connecting rod that is both easily machined in the locations that need machining, and strong in the locations that need strength.

GROUNDS OF REJECTION TO BE REVIEWED ON APPEAL

In the Final Office Action dated September 22, 2005, claims 1, 2, 4, 19 and 21-25 stand variously rejected under 35 U.S.C. §112, first paragraph, 35 U.S.C. §112, second paragraph, and 35 U.S.C. §102(b). As the Advisory Action of February, 2006, provides no relief with respect to any rejection of any claim, it is presumed that the rejections of these claims under each statute section/subsection still stand. Claim 3 is objected to as being dependent from a rejected claim (but would be allowable if placed in independent form).

A brief statement of each ground of rejection follows.

I. Rejections Under 35 U.S.C. §112, First Paragraph

In the Final Office Action, claims 19 and 21-25 were rejected under 35 U.S.C. §112, first paragraph, as failing to comply with the written description requirement.

Claim 19 is specifically rejected on the grounds that it "claims(s) subject matter which was not described in the specification in such a way as to reasonably convey to one skilled in the relevant art that the inventor at the time the application was filed, had possession of the claimed invention." In support of this rejection, the Office Action alleges that the ***drawings*** do not show features such as "(a) the lowest fatigue strength portion in at least one of the big and small ends 20 and 60," and "(b) the "variable fatigue strength portion in each of the first and second joining sections 30 and 50 and the connecting beam section 40." The Office Action further alleges that "the ***drawings*** do not show any variable fatigue strength portion Q." (September Office Action, paragraph spanning pages 4-5, emphasis added.)

The Office Action further alleges, under the auspices of claiming subject matter which was not described in the specification in such a way as to reasonably convey possession, that it "***is unclear as to how Appellant makes***: (a) the lowest fatigue strength portion in at least one of the big and small ends 20 and 60"; and (b) the "claimed variable fatigue strength portion in the sections 30, 40 and 50." (September Office Action, page 5, first full paragraph,

EXHIBIT APPENDIX III

The attached document was utilized as a reference against Appellants for the first time in the Final Office Action of September, 2005, which is where it first appears in the record.

emphasis added.)

II. Rejections Under 35 U.S.C. §112, Second Paragraph

In the Final Office Action, claims 19 and 21-25 were rejected under 35 U.S.C. §112, second paragraph, as being indefinite for failing to particularly point out and distinctly claim the subject matter which Appellant regards as the invention.

Specifically, the Office Action asserts that it "is unclear which portions of the connecting rod are the lowest fatigue strength portion and the variable fatigue strength portion claimed in claim 19. Appellant is respectfully urged to ***identify each claimed element with reference to the drawings***." (September Office Action, page 5, third full paragraph, emphasis added.)

Further, the Office Action asserted that the recitations following the second "wherein" clause in claim 19 is not "***understood since the drawings***, such as, Figs. 1, 2, 11, 20, and 21 do not show the instant claimed features." (September Office Action, page 5, fourth full paragraph, emphasis added.)

III. Rejections Under 35 U.S.C. § 102

In the Office Action of September, 2005, Claims 1, 2 and 4 stand rejected under 35 U.S.C. §102(b) as being anticipated by Japanese Utility Model JP 10-306317 ("JP '317"). Claim 1 is further rejected under this same statute in view of U.S. Patent No. 5,048,368 to Mrdjenovich ("Mrdjenovich") or in view of U.S. Patent No. 5,737,976 to Haman ("Haman").

In rejecting these claims, the Office Action disregards the last recitation of claim 1 (regarding the strength distribution) as an ***inherent*** feature of the prior recitations of the claim, and asserts that the subject matter of this disregarded recitation is ***inherent*** in the prior art.

ARGUMENT

Each ground of rejection is traversed for the following reasons.

I. Rejections of Claims 19 and 21-25 Under 35 U.S.C. §112, First Paragraph

In view of the fact that the September 2005 Final Office Action only contains comments directed to independent claim 19 (from which claims 21-25 depend), Appellants hereby argue against the rejection of claims 19 and 21-25 under 35 U.S.C. §112, first paragraph, for failure to comply with the written description requirement as a group, *i.e.*, based only on the wording of independent claim 19.

A. Preliminary Matter: The language of claim 19 is supported by the *original application as filed*, because claim 19 claim is almost a verbatim copy of both claim 19 as originally filed and a paragraph in the specification, the only difference being that claim 19 was amended to address possible antecedent basis issues.

Appellants traverse the allegation of failure to comply with the written description requirement, ***relying on (i) the fact that an originally filed claim provides its own support vis-à-vis the written description requirement, and (ii) the fact that language of the specification may be relied on to support a claim with respect to the written description requirement***[1]. In this regard, Appellants provide below a reproduction of claim 19 showing how it differs only slightly from originally filed claim 19, followed by a claim chart comparing claim 19 as pending to the language of the paragraph spanning pages 3-4 (page 3, line 15 to page 4, line 12) of the specification as originally filed.

As may be seen, the only difference between claim 19 as pending, and claim 19 as originally filed and the specification as originally filed is that certain ***portions*** of the connecting rod were provided with a *name*, and a singular/plural antecedent basis issue

[1] Applicants note that the PTO does not allege that the ***amendments*** to claim 19 run afoul of the written description requirement; the Office Action's only basis its rejection for lack of written description support on the allegation of deficiencies in the drawings and the allegation that it is unclear as to how to make the connecting rod of claim 19 (both of which are unfounded), as will be detailed below. Regardless, Applicants hereby undertake an analysis of the amendments to claim 19 to remove all possible doubt as to whether claim 19 has support in the originally filed application.

(which had resulted in a previous indefiniteness rejection) was alleviated. That is, Appellants have only amended claim 19 to alleviate present and future possible antecedent basis issues.

Specifically, instead of reciting "a *portion* which is the smallest in cross sectional area," the claim was amended to recite "a **smallest cross sectional area** *portion* which is the smallest in cross sectional area." That is, "a *portion* which is the smallest in cross sectional area" was simply given a *name*: "a smallest cross sectional area *portion*." The same is the case with respect to "a lowest fatigue strength *portion*," and "a variable fatigue strength *portion*." Finally, the term "sections" was amended to "section" to overcome the antecedent basis rejection present in the prior Office Action (the Office Action of March, 2005).

19. (As compared to originally filed claim 19) A high-strength connecting rod comprising:

a connecting beam section serving as a main body of the connecting rod, the connecting beam section having a **smallest cross sectional area** portion which is the smallest in cross sectional area throughout the connecting rod;

a big end located at a first end side of the connecting beam section;

a small end located at a second end side of the connecting beam section, the second end side being axially opposite to the first end side;

a first joining section located between the connecting beam section and the big end to connect the connecting beam section and the big end; and

a second joining section located between the connecting beam section and the small end to connect the connecting beam section and the small end;

wherein each of the first and second joining sections gradually and continuously decreases in cross sectional area toward the connecting beam section;

wherein a **lowest fatigue strength** portion which is the lowest in fatigue strength exists in at least one of the big and small ends, and a **variable fatigue strength** portion which varies in fatigue strength exists in each of the first and second joining sections and in the connecting beam ~~sections~~ **section**;

wherein a product of the cross sectional area and the fatigue strength at a cross section of each of the joining and connecting beam ~~sections~~ section is equal to or greater than a product of the cross sectional area and the fatigue strength in the smallest cross sectional area portion in the connecting beam section.

Claim 19 as Pending:	**Specification As Originally Filed, Page 3, line 15 to page 4, line 12:**
A high-strength connecting rod comprising:	A third aspect of the present invention resides in a high-strength connecting rod comprising
a connecting beam section serving as a main body of the connecting rod, the connecting beam section having a smallest cross sectional area portion which is the smallest in cross sectional area throughout the connecting rod;	a connecting beam section serving as a main body of the connecting rod, the connecting beam section having a portion which is the smallest in cross sectional area throughout the connecting rod.
a big end located at a first end side of the connecting beam section;	A big end is located at a first end side of the connecting beam section.
a small end located at a second end side of the connecting beam section, the second end side being axially opposite to the first end side;	A small end is located at a second end side of the connecting beam section, the second end side being axially opposite to the first end side.
a first joining section located between the connecting beam section and the big end to connect the connecting beam section and the big end; and	A first joining section is located between the connecting beam section and the big end to connect the connecting beam section and the big end.
a second joining section located between the connecting	A second joining section is located between the connecting beam section and

WASH_1593266.3

beam section and the small end to connect the connecting beam section and the small end;	the small end to connect the connecting beam section and the small end.
wherein each of the first and second joining sections gradually and continuously decreases in cross sectional area toward the connecting beam section;	In this connecting rod, each of the first and second joining sections gradually and continuously decreases in cross sectional area toward the connecting beam section.
wherein a lowest fatigue strength portion which is the lowest in fatigue strength exists in at least one of the big and small ends, and a variable fatigue strength portion which varies in fatigue strength exists in each of the first and second joining sections and in the connecting beam section;	Additionally, a portion which is the lowest in fatigue strength exists in at least one of the big and small ends, and a portion which varies in fatigue strength exists in each of the first and second joining sections and in the connecting beam sections.
wherein a product of the cross sectional area and the fatigue strength at a cross section of each of the joining and connecting beam section is equal to or greater than a product of the cross sectional area and the fatigue strength in the smallest cross sectional area portion in the connecting beam section.	Further, a product of the cross sectional area and the fatigue strength at a cross section of each of the joining and connecting beam sections is equal to or greater than a product of the cross sectional area and the fatigue strength in the smallest cross sectional area portion in the connecting beam section.

As may be clearly seen from the above, there was irrefutable written description support for current claim 19 at the time that the application was filed. Apart from a few grammatical differences, originally filed claim 19 and the language of the paragraph spanning pages 3 to 4 of the specification is an exact duplicate of the language of the claim. Clearly, then, the skilled artisan would have recognized that Appellants had possession of the claimed subject matter.

Further, MPEP §2106(V)(B), entitled "Determining Whether the Claimed Invention Complies with 35 U.S.C. §112, First Paragraph Requirements," subsection 1, states, immediately after discussing the "reasonable conveyance" requirement (used as a basis to reject claim 19 and its dependencies) that the "claimed invention subject matter *need not be described literally, i.e., using the same terms*, in order for the disclosure to satisfy the description requirement." (Emphasis added) Appellants respectfully submit that the claims of the present invention find sufficient written description in the as-filed specification.

B. **<u>First Allegation in the Office Action Regarding Written Description</u>:** The Office Action alleges, to support the rejection of claim 19 as failing to comply with the written description requirement, that "the ***drawings*** do not show the claimed features such as "(a) the lowest fatigue strength portion in at least one of the big and small ends 20 and 60." (September Office Action, paragraph spanning pages 4-5, emphasis added.)

Notwithstanding the fact that claim 19 has almost verbatim support in the originally filed application, as detailed above, Appellants further submit that failure to show a claimed feature in the drawings of an application (assuming *arguendo* that this is the case), in view of an adequate originally filed claim and/or an adequate specification text, does not and cannot result in a written description deficiency, as the specification and/or claims, without the drawings, may provide written description for a claim. (For that matter, an adequate text without any drawings may also provide enablement as well.)

Further, "the lowest fatigue strength portion" is, at least in part, governed by material property, and thus there is no way to schematically show this feature. By way of analogy, a piece of 304 stainless steel heat treated to 150 KSI would look the same even if it was instead heat treated to 170 KSI (the former being of *lower* fatigue strength), at least in regards to the ink drawn schematics utilized in a patent application. It is submitted that the ordinary artisan would not need a schematic representation of "a lowest fatigue strength portion" to recognize that Appellants were in possession of the invention as claimed in claim 19 – the specification and originally filed claim 19 proving such evidence of possession.

Appellants respectfully submit that the MPEP supports the above positions taken by Appellants in traversing the rejection of claim 19 under 35 U.S.C. §112, first paragraph. For example, Appellants point to MPEP §2163.04(i) entitled "***Burden on the Examiner*** with Regard to the Written Description Requirement," (emphasis added) which states that the

> inquiry into whether the description requirement is met must be determined on a case-by-case basis and is a question of fact. *In re Wertheim*, 541 F.2d 257, 262, 191 USPQ 90, 96 (CCPA 1976). A description as filed is presumed to be adequate, unless or until sufficient evidence or reasoning to the contrary has been presented by the examiner to rebut the presumption. See, e.g., *In re Marzocchi*, 439 F.2d 220, 224, 169 USPQ 367, 370 (CCPA 1971). The examiner, therefore, must have a reasonable basis to challenge the adequacy of the written description. The examiner has the initial burden of presenting by a preponderance of evidence why a person skilled in the art would not recognize in an Appellant's disclosure a description of the invention defined by the claims. *Wertheim*, 541 F.2d at 263, 191 USPQ at 97.

It is respectfully submitted that no evidence has yet been proffered by the PTO to support the rejection of claim 19 under 35 U.S.C. §112, first paragraph. Appellants provide further excerpts from this MPEP section in Exhibit Appendix I in support of their position, and respectfully submit that the requirements of the MPEP vis-à-vis establishing a *prima facie* case of a lack of written description have not been established.

C. Second Allegation in the Office Action Regarding Written Description: The Office Action alleges that "the ***drawings*** do not show the claimed features such as, . . . (b) the "variable fatigue strength portion in each of the first and second joining sections 30 and 50 and the connecting beam section 40." (Final Office Action, paragraph spanning pages 4-5, emphasis added.)

In response, Appellants refer to the above discussions in section "B" regarding this allegation, as this allegation is not valid with respect to the written description requirement for the same reasons detailed with respect to the allegation that the drawings do not show the lowest fatigue strength portion.

D. Third Allegation in the Office Action Regarding Written Description: The Office Action alleges that it "is unclear as to how Appellant makes: (a) the claimed lowest fatigue strength portion in at least one of the big and small ends 20 and 60; and (b) the claimed variable fatigue strength portion in the sections 30, 40 and 50." (September Office Action, page 5, first full paragraph.)

First, assuming *arguendo* that the factual premise in the Final Office Action were true, this has no bearing on whether a claim fails the written description requirement in view of an adequate originally filed specification. Second, Appellants respectfully submit that the ordinary artisan would have understood how to make the claimed invention based on the specification as originally filed. Again, the presence or absence of drawings, in and of itself, does not govern the question of whether the skilled artisan would have understood how to make an invention . (It further has nothing to do with the written description requirement of a claim containing language that is almost exactly verbatim from an originally filed specification.)

E. Allegation Made During the Interview of January 18, 2006, Regarding Written Description: During the interview of January 18, 2006, the PTO alleged that there is no support for claim 19 because the specification uses the term "or" and claim 19 uses the phrase "at least one of" with respect to the big and small ends.[2] In response, Appellants traverse this assertion, ***relying on the fact that an originally filed claim provides its own support vis-à-vis the written description requirement***. (Interview Summary of January 18, 2006). In this regard, claim 19, as originally filed, uses the language "*at least one of*," [3]and thus there is written description support for claim 19 in the originally filed application.

---

[2] Specifically, it was alleged that there is no support for claim 19 because the specification, at page 24, lines 11-14, states that the "connecting rod has a portion of the smallest cross sectional area in its connecting beam section, a portion of the lowest fatigue strength at its big ***or*** small end," as compared to the recitation of "a lowest fatigue strength portion which is the lowest in fatigue strength exists **in at least one of** the big and small ends, and a variable fatigue strength portion which varies in fatigue strength exists in each of the first and second joining sections and in the connecting beam section."

[3] Claim 19, as originally filed, recites "wherein a portion which is the lowest in fatigue strength exists in *at least one of* the big and small ends, and a portion which varies in fatigue strength exists in each of the first and second joining sections and in the connecting beam sections." (Emphasis added.)

Further, as seen above, the paragraph spanning pages 3-4 also utilizes the language "at least one of" with respect to the big and small ends. Thus, the specification provides support for this claim as well.

However, in order to advance prosecution, and without prejudice or disclaimer, Appellants proffered an amendment to the specification, as seen above, to recite on page 24 the exact language at issue of claim 19. No new matter has been added in view of originally filed claim 19 and specification at pages 3-4. This amendment was refused entry on the grounds that the preamble "In another embodiment" "raises new issues that would require further consideration." Although Appellants disagree, and submit that no new issues are raised by this amendment, they likewise do not believe that the above proffered amendment to page 24 is necessary, as the specification and originally filed claim 19 support the claim language "in at least one of" the big and small ends.

In sum, claims 19 and 21-25 are supported by the specification as originally filed. These claims do not fail the written description requirement, and the absence of claimed subject matter in the drawings is not dispositive of this issue. Reversal of the rejections under 35 U.S.C. §112, first paragraph, is respectfully requested.

II. <u>Rejections Under 35 U.S.C. §112, Second Paragraph</u>

In the Final Office Action, claims 19 and 21-25 were rejected under 35 U.S.C. §112, second paragraph, as being indefinite for failing to particularly point out and distinctly claim the subject matter which Appellant regards as the invention. The basis of the rejection appears to be the alleged lack of a depiction of various claim features in the ***drawings***.

A. <u>First Allegation in the Office Action Regarding Indefiniteness</u>: The September 2005 Office Action asserted that it "is unclear which portions of the connecting rod are the lowest fatigue strength portion and the variable fatigue strength portion claimed in

claim 19. Appellant is respectfully urged to identify each claimed element with reference to the ***drawings***." (Page 5, third full paragraph, emphasis added.)

Appellants respectfully submit that, assuming *arguendo* there is in fact a lack of such depiction, this does not make the claims indefinite. The claims clearly define the invention. Claim 19 reads:

> wherein a ***lowest fatigue strength portion which is the lowest in fatigue strength*** exists in at least one of the big and small ends, and ***a variable fatigue strength portion which varies in fatigue strength exists*** in each of the first and second joining sections and in the connecting beam section.

(Emphasis added.) Appellants submit that the claimed portions are self-defining, and thus it is clear which portions of the connecting rod are which. Although the claims may be broad, this fact alone does not make them indefinite. (MPEP §2173.04)

Appellants respectfully remind the PTO that claims are to be evaluated with the ordinary skill test: "Acceptability of the claim language depends on whether one of ordinary skill in the art would understand what is claimed, in light of the specification." (MPEP §2173.05(b).) Appellants respectfully submit that one of ordinary skill would readily understand claims 19 and 21-25, and no evidence has been proffered to the contrary. Reconsideration is requested.

B. <u>Second Allegation in the Office Action Regarding Indefiniteness</u>: The September 2005 Office Action asserts that the recitations following the second "wherein" clause in claim 19 (see above chart) are not "understood since the ***drawings***, such as, Figs. 1, 2, 11, 20, and 21 ***do not show the instant claimed features***." (Page 5, fourth full paragraph, emphasis added.)

In response, Appellants again submit that the drawings are not definitive of whether a claim is indefinite, and submit that the recitations of claim 19 at issue in this allegation are not indefinite for the pertinent reasons just detailed above in section "A."

In sum, claim 19 and 21-25 are not indefinite. Reversal of this rejection is respectfully requested.

III. Rejections Under 35 U.S.C. § 102

There are only two issues that need be resolved to determine whether the claims are anticipated in view of the prior art, both of which are interrelated: **(1)** is the Office Action correct, both procedurally and factually, to ***disregard*** a recitation in claim 1 on the grounds that the recitation only recites *inherent* features of previous claim elements (Appellants say no, it is not an inherent feature), and **(2)** is the Office Action correct, both procedurally and factually, in its assertion that the disregarded recitation is ***inherently*** present in the cited prior art references (to which Appellants also say no)?

The specifics of the rejections will now be discussed.

In the Office Action of September, 2005, Claims 1, 2 and 4 stand rejected under 35 U.S.C. §102(b) as being anticipated by Japanese Utility Model JP 10-306317 ("JP '317"). Claim 1 (but no other) is further rejected under this same statute in view of U.S. Patent No. 5,048,368 to Mrdjenovich ("Mrdjenovich") or in view of U.S. Patent No. 5,737,976 to Haman ("Haman").

MPEP § 2131, entitled "Anticipation – Application of 35 U.S.C. 102(a), (b), and (e)," states that a "claim is anticipated only if each and every element as set forth in the claim is found, either expressly or inherently described, in a single prior art reference." Section 103 amplifies the meaning of this anticipation standard by pointing out that anticipation requires that the claimed subject matter must be "*identically* disclosed or described" by the prior art reference. (Emphasis added.) It is respectfully submitted that the cited references do not describe (expressly or inherently) each and every element of independent claim 1, and thus also does not describe the subject matter of the claims that depend therefrom.

A. <u>Claim Element that is Missing from all the References is not an Inherent Feature of the Other Claim Elements</u>: Claim 1 recites that "each of the first and second ***joining sections*** gradually and continuously decreases in cross sectional area toward the connecting beam section and ***has a strength distribution in which a strength increases with a decrease in the cross sectional area.***" (Emphasis added.) That is, there are two portions of the connecting rod where, as there is *less and less* material present, the strength of the connecting rod *increases*, in contrast to how strength usually *decreases* with *less and less* material present. In an exemplary embodiment according to claim 1, this strength distribution is achieved by "controlling hardening by heat treatment (hardening), so that a distribution may be produced in the hardening temperature and/or tempering time during the quenching of the joining sections." (Specification, page 12, lines 21-26.) The strength distribution is thus obtained as a result of different material properties at various locations of the connecting rod. The claim does link the strength distribution to the geometry of the joining sections, because this is where the claimed phenomenon takes place, to differentiate the invention from the prior art, but it is a change in material properties along the connecting rod that gives rise to the strength distribution.

The Final Office Action is not entirely clear as to how it treats this recitation. While first asserting that the cited references teach each element of claim 1, the Office Action goes on to assert that the above-quoted recitation is an ***inherent result*** of the other features of claim 1 (Final Office Action, paragraph spanning pages 6-7, and pages 10 and 11), thus raising ambiguity as to whether the Office Action gives patentable weight to this recitation.

Appellants therefore first address the fact that the patentably distinct language following the "wherein" clause is a feature of the invention that must be given patentable weight, as it is impossible to evaluate the prior art without first determining the scope of the claims. (The facts that relate to why the strength distribution recitation is not an inherent feature of the other elements of claim 1 also relate to why the claimed distribution is not inherently present in the prior art, and thus the two questions are intertwined.)

In apparent support of its action to disregard the above-quoted language, the Office Action cites *Texas Instruments v. ITC*,[4] and *Griffin v. Bertina*[5] as standing for the position that "the 'wherein' clause or 'whereby' clause that merely expresses an inherent result, adds nothing to claim's patentability." (Final Office Action, page 6, paragraph spanning pages 6 and 7.) *Texas Instruments* dealt with the clause "***whereby***," not "***wherein***." *Griffin* did deal with the term "wherein" (albeit in the context of establishing a reduction to practice date in an interference), but held that the wherein clause ***could <u>not</u>*** be disregarded. Specifically, the court in *Griffin* stated that they were

> not persuaded by Griffin's arguments that the "wherein" clauses merely state the inherent result of performing the manipulative steps. A party seeking to show that it need not establish reduction to practice of every feature recited in the count based on the alleged inherency of some of those features must show that the "inherent" properties add nothing to the count beyond the other recited limitations and are not material to the patentability of the invention.

Griffin V. Bertina, 285 F.3d 1029, 62 USPQ2d 1431, 1434 (Fed. Cir. 2002).

Moreover, even if these case were to be abstracted to cover the present situation, the result of *Texas Instruments* would not apply, and the result of Griffin would apply (the strength distribution recitation would be considered a recitation), because the above-quoted language does more than merely express an inherent result. That is, the case law cited in the Office Action only supports a decision that each claim recitation in claim 1 must be given patentable weight.

The key issue in *Texas Instruments* and *Griffin* was the determination of whether a claimed feature was an ***inherent or necessary*** result of the other claim recitations. "Inherent"

---

[4] *Texas Instruments v. International Trade Commission,* 988 F.2d 1165, 26 USPQ2d 1018 (Fed. Cir. 1993).

[5] *Griffin V. Bertina,* 285 F.3d 1029, 62 USPQ2d 1431 (Fed. Cir. 2002).

means that ***a feature is necessarily present***. (See MPEP § 2112 – discussed in greater detail below with respect to the prior art.) That is, the feature is always present.

The above-quoted recitation regarding the strength distribution is not ***necessarily present*** in the prior recited elements of claim 1. This is ***evinced by the fact that in prior art connecting rods, strength <u>decreases</u> with a decrease in cross sectional area***, and thus an increase in strength with decreasing cross section is not ***always*** present. That is, the strength distribution clause of claim 1 does more than merely express an inherent result. There is simply nothing in the recitations preceding this language that ***require*** (or even suggest, for that matter) that "the first and second joining sections gradually and continuously decrease in cross sectional area toward the connecting beam section" have "a strength distribution in which a strength increases with a decrease in the cross sectional area," as would be necessary for a result to be ***inherent***.

The PTO neither cites evidence nor sets forth any explanation to support its position that the above quoted language, following the "wherein" clause of claim 1, is an inherent result of the prior recitations. In fact, aside from misapplying two cases (*Texas Instruments* and *Griffin*), one of which (*Griffin*) supports giving patentable weight to the recitations at issue, the only other arguments apparently made in support of this position are circular, asserting simply that because claim 1 recites certain features, the other feature is inherent. (This is discussed in greater detail at the end of this section.) Appellants submit that, for the PTO to maintain its assertion of inherency, the PTO must identify which element of claim 1 results in the inherency of the above-quoted language **and explain why that element results in a strength distribution that increases with decreasing cross sectional area, instead of decreasing with decreasing cross sectional area**

Appellants respectfully submit that the claimed elements, individually or combined in any combination, do not inherently result in ***a strength distribution in which a strength increases with a decrease in the cross sectional area***, and the PTO has not proffered evidence to the contrary.

Indeed, Appellants submit just the opposite: that the elements of claim 1 prior to the above-quoted language, *without more*, do not result in joining sections having a strength distribution in which a strength increases with a decrease in the cross sectional area. Appellants "more" is obtained as detailed in the specification (controlled martensitic transformation, *etc.*) and differentiates the invention of claim 1 from the prior art. The recited strength distribution is not an inherent feature of the prior recited elements, and, in fact, is opposite of what is usually the case (the strength distribution is one that usually decreases with a decrease in the cross sectional area). This is the case not only with the preceding elements of claim 1, but the connecting rods of the prior art as well, as will be detailed below.

The paragraph spanning pages 10-11 of the Final Office Action states that it is known that "the strength of material improves by the hardening, heat-treatment, and cold forging process, *etc.*" citing the text "*Mechanical Design and Systems Handbook.*" Appellants do not deny this is the case with respect to most steels. The Office Action goes on to assert that "Appellant uses the methods such as hardening, heat-treatment, and cold forging process." Appellants do not deny the use of heat-treatment in obtaining embodiments of their connecting rod, ***although the use of heat-treatment is not recited in claim 1***. The Office Action then comes to the conclusion that "[t]herefore, the strength distribution of Appellant's connecting rod ***must*** be improved particularly in the areas where the cross sectional [sic] is decreased to prolong life of the connecting rod as taught by standard text books." (Emphasis added.) That is, because Appellants utilize a process that is ***not recited*** in claim 1, the results of that process must necessarily be present in / flow from claim recitations that have nothing to do with the unclaimed process (which of the above 6 claim elements is related to heat-treatment?). Because heat-treatment is not claimed in claim 1, this argument further underscores the fact that the recited strength distribution is not inherently present in the preceding claim elements.

Also, there is nothing about the preceding facts (heat-treatment is known to improve the strength of most steels, and Appellants use heat-treatment to prepare embodiments of their connecting rod) that results in a scenario where "the strength distribution of Appellant's

connecting rod ***must*** be improved particularly in the areas where the" cross sectional area is decreasing. This evinces, at most, only the capability of "improving" strength. Further, even if the alleged "must" result is the case, it still does not result in a strength distribution that increases with decreasing cross sectional area. Instead, it results in, *arguendo,* a strength distribution that is "improved" in the areas where cross sectional area is decreasing. "Improvement" is not the same as the strength distribution as claimed. Indeed, these arguments presented in the Office Action further underscore the fact that the strength distribution recitation is decidedly not inherent in the preceding claim recitations. It is respectfully submitted that it is wrong that the Office Action does not view the claimed strength distribution as giving patentable weight to the invention, as the claimed strength distribution is not an ***inherent*** feature of any of the preceding claims, and, in fact, differentiates the present invention from the prior art.

In sum, the recitation of ***a strength distribution in which a strength increases with a decrease in the cross sectional area*** is not an inherent feature of the previous claimed elements in claim 1 (nor is it an inherent feature of the prior art, as will be discussed below), and thus this recitation must be given patentable weight.

B. **<u>First Allegation of Anticipation</u>:** The September Office Action incorrectly alleges that claims 1, 2 and 4 are anticipated under 35 U.S.C. §102(b) by JP '317.

1. All Claims

JP '317 teaches a connecting rod. Assuming *arguendo* that JP '317 meets each and every recitation of claim 1 prior to the strength distribution recitation, JP '317 does not teach, either expressly or inherently, first and second joining sections that gradually and continuously decrease in cross sectional area as recited ***and have a strength distribution in which a strength <u>increases</u> with a <u>decrease</u> in the cross sectional area.***

The Office Action asserts on page 6 that JP '317 teaches each element of claim 1, reciting all of claim 1, word for word. After each recited element, the Office Action identifies

where the element may be found in the art, ***with the exception of the recitation regarding the strength distribution***. This is because JP '317 does not teach this distribution.

There is no teaching, either expressly or inherently, in JP '317, that the alleged ***joining sections*** have a strength distribution in which a strength increases with a decrease in the cross sectional area, where ***the joining sections are respectively located between*** the connecting beam section and the big end to connect the connecting beam section and the big end, and located between the connecting beam section and the small end to connect the connecting beam section and the small end, as is recited in claim 1. No evidence is proffered by the PTO to the contrary.

Various cross sections are depicted in Figure 12 of JP '317, and there are numerical values associated with various positions within the cross sections. It is unclear what these numerical values mean, but assuming *arguendo* that these numerical values are related to "strength," JP '317 still does not anticipate claim 1. This is at least because no cross section is present in any ***joining section*** of JP '317. Instead, cross sections "A" and "C" are taken through the big end and the little end, and cross section "B" is taken through the connecting portion. These are not through any joining section, and thus, to the extent that JP '317 teaches features of the material properties at these locations (assumed *arguendo* to be the case), JP '317 still does not teach, either expressly or inherently, features regarding the strength distribution in any joining sections.

It is unclear whether the Office Action relies on an inherency argument to remedy the above identified deficiencies of JP '317 or whether the inherency argument is used (incorrectly) to simply eviscerate the strength distribution recitation from the claim, as detailed in section "A" above. To the extent that the Office Action is asserting that the prior art inherently has the claimed strength distribution, Appellants traverse such assertion, pointing to MPEP §2112, which states that while "a rejection under 35 U.S.C. §102/103 can be made when the prior art product seems to be identical except that the prior art is silent to an inherent characteristic," the "[E]xaminer ***must*** provide ***rationale or evidence*** tending to

show inherency." (MPEP § 2112, subsections 3 and 4, emphasis added.) It is respectfully submitted that no evidence tending to show inherency has been provided in the present Office Action. Further, in considering how the inherency concept is being used in the Office Action, it is respectfully submitted that § 2112 inherency is not being properly implemented. In arriving at this conclusion, Appellants rely on the following excerpt from MPEP § 2112:

> The fact that a certain result or characteristic may occur or be present in the prior art is not sufficient to establish the inherency of that result or characteristic. *In re Rijkaert*, 9 F.3d 1531, 1534, 28 USPQ2d 1955, 1957 (Fed. Cir. 1993) (reversed rejection because inherency was based on what would result due to optimization of conditions, not what was necessarily present in the prior art); *In re Oelrich*, 666 F.2d 578, 581-82, 212 USPQ 323, 326 (CCPA 1981). "To establish inherency, the extrinsic evidence 'must make clear that ***<u>the missing descriptive matter is necessarily present</u>*** in the thing described in the reference, and that it would be so recognized by persons of ordinary skill. Inherency, however, may not be established by probabilities or possibilities. The mere fact that a certain thing may result from a given set of circumstances is not sufficient.'" *In re Robertson*, 169 F.3d 743, 745, 49 USPQ2d 1949, 1950-51 (Fed. Cir. 1999) (citations omitted) (The claims were drawn to a disposable diaper having three fastening elements. The reference disclosed two fastening elements that could perform the same function as the three fastening elements in the claims. The court construed the claims to require three separate elements and held that the reference did not disclose a separate third fastening element, either expressly or inherently.)

(Emphasis added.) Inherency means that *the missing descriptive matter is necessarily present* in the reference. The courts have allowed the PTO to rely on inherency arguments to free the PTO from the necessity of finding references which explicitly state that inherent elements are present. This is because certain characteristics are inherent, the references will most probably not mention these elements, and, as such, will be difficult to find. For example, it is not necessary to find a reference that explicitly states that plutonium 239 is radioactive, as plutonium 239 is always radioactive. That is, radioactivity is an inherent feature of plutonium 239. However, inherency is not a panacea that enables the PTO to use references which are *deficient* in teaching certain elements of a claim. Recognizing the power of the inherency argument, the courts have tempered its use, as is seen in § 2112, where the PTO has

stipulated that examiners must follow certain procedures before invoking inherency: the "examiner must provide rationale or evidence tending to show inherency." In the present case, no such rationale or evidence has been provided in the Office Action. Just because it may be ***desirable or useful*** to have a connecting rod having a strength distribution as claimed does not mean that such properties are ***always*** present. Just the opposite is true: as pointed out above, connecting rods typically have strength distributions that decrease with decreasing cross sectional area. The subject matter claimed in claim 1 is not ***necessarily present*** in the references. It is entirely probable that the references will be practiced without a strength distribution that changes as claimed. Just as was the case of the third fastener in the example provided in the MPEP quoted above, the subject matter of Appellants' claims is not expressly or inherently disclosed in any of the cited references.

With the above in mind, the Office Action asserts on the second full paragraph on page 11 of the Final Office Action that

> the connecting rod of JP '317, Mrdjenovich, and Haman have joining sections that are gradually and continuously decreased in cross sectional area towards a connecting beam. Thus, the connecting rods of JP '317, Mrdjenovich, and Haman are expected to behave in the same manner as Appellant's connecting rod because they all have the same sectional profiles.

That is, the Office Action asserts that because the connecting rods of the prior art look like they fall within the scope of ***some*** of the recitations of claim 1, the ***other*** recitations are met by the prior art. This is not the standard for rejecting a claim as anticipated, and to the extent that the Office Action is asserting an inherent result or characteristic, insufficient evidence / rationale has been proffered.

Still further, as demonstrated above in section "A," the strength distribution is not an inherent feature of any of the preceding elements of claim 1, and, therefore, to the extent that corresponding elements may be found in a prior art reference, it necessarily follows that, without more, those corresponding elements likewise do not have as an inherent feature the strength distribution as claimed.

To reiterate, as detailed in the specification (and above), Appellants obtain their strength distribution by the way portions of the connecting rod are treated – some portions are treated differently from other portions of the connecting rod during the manufacturing process to obtain different material properties (*see, for example,* Fig. 9 in view of Fig. 1). The claimed strength distribution is thus obtained as a result of different material properties at various locations of the connecting rod. By linking the strength distribution to the geometry of the joining sections, Appellants differentiate the apparatus invention from the prior art, but it is a change in material properties along the different sections of the connecting rod that gives rise to the claimed strength distribution. Accordingly, two connecting rods that may have the exact same dimensions may also have different structural characteristics due to the strength of the material. Thus, even though schematics of the prior art may look like they fall within the scope of some of the claim elements of claim 1, the prior art does not inherently teach the remaining elements of claim 1.

In sum, claim 1 is not anticipated because the prior art does not teach, either expressly or inherently, a connecting rod having first and second joining sections that gradually and continuously decrease in cross sectional area toward a connecting beam section and having a strength distribution in which a strength increases with a decrease in the cross sectional area.

2. Claim 2:

Claim 2 recites a connecting rod as claimed in Claim 1, wherein the strength distribution is based on a proportion (%) of martensite. Claim 2 is allowable at least due to its dependency from claim 1, but also because JP '317 does not teach, either expressly or inherently, this feature.

In rejecting claim 2, the Office Action repeats the recitations of claim 2, and asserts that the English abstract of JP '317 teaches these recitations. (See first full paragraph of page 7 of the September Office Action.) While it is true that JP '317 does refer to quenching

to promote martensitic transformation, there is no teaching in JP '317 that the specific strength distribution is based on a proportion (%) of martensite. Thus, the Office Action appears to be making another inherency argument.

Indeed, the Office Action acknowledges that "JP '317 does not ***explicitly*** so describe[e]"the recitations of claim 2. (Final Office Action, end of paragraph spanning pages 11-12, emphasis added). Thus, for JP '317 to anticipate claim 2, it must ***inherently*** teach each element of the claim, as MPEP § 2131 states that a "claim is anticipated only if each and every element as set forth in the claim is found, either ***expressly*** or ***inherently*** described, in a single prior art reference." Therefore because, as the Office Action recognizes, JP '317 does not explicitly describe claim 2, the only way for JP '317 to anticipate claim 2 is for the missing recitations of claim 2 to be inherent in JP '317, which it is not.

As discussed above with respect to claim 1, inherency means that the missing subject matter is necessarily present – it is always there (MPEP §2112). It is not always the case that a strength distribution is controlled by the percentage of martensite in a structure. This is because there are other structures may be present in JP '317 that would influence the strength. For example, the presence of ***bainite*** would influence the strength distribution, and thus it does not necessarily follow that simply because a structure contains martensite, that structure has a strength distribution that changes on a proportion of martensite, as bainite that is present also influences the strength. In support of this proposition, Appellants proffer the attached article in Exhibit Appendix II as highlighted.[6]

In the paragraph spanning pages 11-12 of the Final Office Action, the Office Action asserts that it is known that the presence and amount of Martensite present in a steel component may influence strength. Appellants agree. However, JP '317 does not teach a strength distribution ***which varies as claimed*** is based on a proportion of martensite. JP '317

[6] *Review of the Performance of High Strength Steels Used Offshore*, prepared by Cranfield University, 2003 (e.g., pages 6, 35, 71, 107 and 110), evincing that bainite influences strength.

only teaches that a martensitic transformation takes place. Thus, JP '317 does not teach a variation in strength based on a percentage of martensite.

The Office Action generally points to an entire chapter of a text book, *Mechanical Design and Systems Handbook*, (Exhibit Appendix III), ***without specifying a specific page or range of pages of the 33 pages cited***, alleging that the handbook "clearly teach [sic] that the strength distribution is based on a proportion (%) of martensite." (September Office Action, last sentence of paragraph spanning pages 11-12.) This is not so. All that the handbook teaches is that martensite influences the strength of steel. The handbook says nothing about a strength distribution within a piece of steel being based on a proportion of martensite, and nothing about a distribution that ***varies*** as claimed based on a proportion of martensite. Appellants respectfully submit that the recitations of claim 2 are not sufficiently being examined.

Because JP '317 does not explicitly teach each element of claim 2, as is admitted in the Office Action, and because JP '317 does not inherently teach each element of claim 2, claim 2 is not anticipated for a second reason (the first being due to its dependency).

3. Claim 4:

Claim 4 recites a connecting rod as claimed in Claim 2, wherein the strength distribution is formed based on a distribution in at least one of a hardening temperature and a tempering time for each of the first and second joining sections.

In rejecting claim 4, the Final Office Action asserts that the "strength distribution is ***inherently*** formed based on [replication of the claim language of claim 4]," citing the Derwent English language translation of the abstract of JP '317. (September Office Action, page 7, emphasis added.) The Office Action goes on to again generally rely on chapter 17 *Mechanical Design and Systems Handbook*, this time asserting that an entire chapter (without

specifying a page or a page range within the chapter) "clearly teach [sic] that the strength distribution is inherently or must be based on [the claim langue of claim 4]." (September Office Action, page 12, first full paragraph.)

Appellants respectfully submit that there is nothing in the Derwent English Abstract of JP '317 that demonstrates that the strength distribution is *inherently* formed based on a distribution in at least one of a hardening temperatures and a tempering time for each of the first and second joining sections. First, there is no reference to temperature or time in the Abstract. Second, for a feature to be inherent, that feature must occur each and every time, pursuant to MPEP §2112. Since there is no evidence that this feature occurs each and every time the teachings of JP '317 are implemented, claim 4 is not inherently anticipated by this reference.

Moreover, there is nothing in the cited *handbook* regarding a strength distribution within a piece of steel, and nothing regarding a strength distribution that varies as claimed. Further, strength may be influenced by factors other than hardening temperature and a tempering time, such as, for example, whether the metal was cold worked, the composition of the metal, *etc*. Thus, assuming *arguendo* that a reference may be found that teaches that a strength distribution may be based on hardening temperature and tempering time, such a reference still does not make it inherent (*i.e.,* it is ***always*** the case) that JP '317 teaches a strength distribution that is "formed based on a distribution in at least one of a hardening temperate and a tempering time,." as there are other ways to influence strength, and JP '317 is silent as to time and temperature.

In summary, claim 4 is not anticipated because JP '317 does not inherently teach each element of this claim, in addition to being allowable due to its dependency from claim 1.

C. **Second Allegation of Anticipation:** The Final Office Action alleges that claim 1 is anticipated under 35 U.S.C. §102(b) by Mrdjenovich. Specifically, the Office Action directs the Appellant to see "cross sections shown in Figs. 7 and 6 that gradually and continuously decrease in cross sectional area towards the connecting beam section 11 as seen

in Figs. 2 and 3." (September Office Action, page 7, fourth full paragraph.) The Office Action goes on to assert that the connecting rod of Mrdjenovich is "expected to behave in the same manner as Appellant's connecting rod because [it has] the same sectional profiles." (September Office Action, page 11, second full paragraph.)

As with JP '317, the Office Action never identifies where Mrdjenovich teaches a connecting rod where "each of the first and second ***joining sections*** gradually and continuously decreases in cross sectional area toward the connecting beam section and ***has a strength distribution in which a strength increases with a decrease in the cross sectional area.***" Instead, the Office Action again merely asserts, as it did with JP '317, that because the connecting rods of the prior art look like they fall within the scope of ***some*** of the recitations of claim 1, the ***other*** recitations are met by the prior art. Again, Appellants reiterate that this is not the standard for rejecting a claim as anticipated. Moreover, as this feature is not inherent in the prior recitations of claim 1, there is no reason to believe that this feature is inherently present in any corresponding recitations, as detailed above with respect to JP '317.

Appellants thus traverse the rejection of claim 1 in view of Mrdjenovich, and submit that the same arguments detailed above with respect to the deficiencies of JP '317 are applicable to Mrdjenovich.

D. <u>Third Allegation of Anticipation</u>: The September Office Action alleges that claim 1 is anticipated under 35 U.S.C. §102(b) by Haman. Specifically, the Office Action directs the Appellant to see "cross sections 103 and 105 shown in Fig. 2 that gradually and continuously decrease in cross sectional area towards the connecting beam section 101." (September Office Action, page 7, sixth full paragraph.) The Office Action goes on to assert that the connecting rod of Haman is "expected to behave in the same manner as Appellant's connecting rod because [it has] the same sectional profiles." (September Office Action, page 11, second full paragraph.)

As with JP '317, the Office Action never identifies where Haman teaches a connecting rod where "each of the first and second ***joining sections*** gradually and

continuously decreases in cross sectional area toward the connecting beam section and ***has a strength distribution in which a strength increases with a decrease in the cross sectional area.***" Instead, the Office Action again merely asserts, as it did with JP '317, that because the connecting rods of the prior art look like they fall within the scope of ***some*** of the recitations of claim 1, the ***other*** recitations are met by the prior art. Again, Appellants reiterate that this is not the standard for rejecting a claim as anticipated. Again Appellants also point out that as this feature is not inherent in the prior recitations of claim 1, there is no reason to believe that this feature is inherently present in any corresponding recitations, as detailed above with respect to JP '317.

Appellants thus traverse the rejection of claim 1 in view of Haman, and submit that the same arguments detailed above with respect to the deficiencies of JP '317 are applicable to Haman.

IV. Refusal to Evaluate Claims 19 and 21-25 for Anticipation and Obviousness

The Office Action of September, 2005, refuses to examine claims 19 and 21-25, which were rejected under 35 U.S.C. §112, first and second paragraphs, in view of the prior art.

As detailed above, these claims are neither indefinite, nor in violation of the written description requirement. Moreover, the refusal in the Office Action to examine these claims in view of the prior art is in ***direct contradiction*** with MPEP §2143.03, second paragraph, which states that "claim limitation which is considered indefinite ***cannot*** be disregarded. (MPEP §2143.03, second paragraph, emphasis added.) MPEP §2143.03 also states that if "a claim is subject to more than one interpretation, at least one of which would render the claim unpatentable over the prior art, the examiner ***should*** reject the claim as indefinite . . . and ***should*** reject the claim over the prior art based on the interpretation of the claim that renders the prior art applicable." (MPEP §2143.03, second paragraph, emphasis added.)

In any event. There are no rejections of claims 19 and 21-25 over the prior art.

CONCLUSION

Appellants respectfully request that all rejections be reversed for the reasons set forth above.

Respectfully submitted,

Date May 22, 2006

By ______________________

FOLEY & LARDNER LLP
Customer Number: 22428
Telephone: (202) 295-4747
Facsimile: (202) 672-5399

Martin J. Cosenza
Registration No. 48,892

Richard L. Schwaab
Registration No. 25,479

CLAIMS APPENDIX

1. (Original) A connecting rod comprising:

a connecting beam section serving as a main body of the connecting rod;

a big end located at a first end side of the connecting beam section;

a small end located at a second end side of the connecting beam section, the second end side being axially opposite to the first end side;

a first joining section located between the connecting beam section and the big end to connect the connecting beam section and the big end; and

a second joining section located between the connecting beam section and the small end to connect the connecting beam section and the small end;

wherein each of the first and second joining sections gradually and continuously decreases in cross sectional area toward the connecting beam section and has a strength distribution in which a strength increases with a decrease in the cross sectional area.

2. (Original) A connecting rod as claimed in Claim 1, wherein the strength distribution is based on a proportion (%) of martensite.

3. (Original) A connecting rod as claimed in Claim 2, wherein the proportion of martensite (%) changes based on a change of the cross sectional area of each of the first and second joining sections in a manner to satisfy a relationship represented by the following formula:

$$D/D_{min} \geq 1/((1-\alpha) \times Ms/100 + \alpha)$$

where D_{min} is the minimum value of the cross sectional area of each of the first and second joining sections; and α is a value obtained by dividing a buckling stress without hardening by a buckling stress with hardening.

4. (Original) A connecting rod as claimed in Claim 2, wherein the strength distribution is formed based on a distribution in at least one of a hardening temperature and a tempering time for each of the first and second joining sections.

5. (Withdrawn) A connecting rod as claimed in Claim 1, wherein the strength distribution is formed based on a strain introduced into each of the first and second joining sections by a cold forging.

6. (Withdrawn) A connecting rod as claimed in Claim 5, wherein the strain gradually and continuously changes with a change in the cross sectional area of each of the first and second joining sections.

7. (Withdrawn) A connecting rod as claimed in Claim 5, wherein the strain is adjusted in accordance with a dispersion in thickness of a roughly made connecting rod as a material of the connecting rod.

8. (Withdrawn) A connecting rod as claimed in Claim 5, wherein each of the first and second joining sections is subjected to an aging after the cold forging.

9. (Withdrawn) A method of producing a connecting rod including

a connecting beam section serving as a main body of the connecting rod;

a big end located at a first end side of the connecting beam section;

a small end located at a second end side of the connecting beam section, the second end side being axially opposite to the first end side;

a first joining section located between the connecting beam section and the big end to connect the connecting beam section and the big end; and

a second joining section located between the connecting beam section and the small end to connect the connecting beam section and the small end,

the producing method comprising:

gradually and continuously decreasing each of the first and second joining sections in cross sectional area toward the connecting beam section; and

providing to each of the first and second joining sections a strength distribution in which a strength increases with a decrease in the cross sectional area.

10. (Withdrawn) A producing method as claimed in Claim 9, wherein the strength distribution is based on a proportion (%) of martensite.

11. (Withdrawn) A producing method as claimed in Claim 10, wherein the proportion of martensite (%) changes based on a change of the cross sectional area of each of the first and second joining sections in a manner to satisfy a relationship represented by the following formula:

$$D/D_{min} \geq 1/((1-\alpha) \times Ms/100 + \alpha)$$

where D_{min} is the minimum value of the cross sectional area of each of the first and second joining sections; and α is a value obtained by dividing a buckling stress without hardening by a buckling stress with hardening.

12. (Withdrawn) A producing method as claimed in Claim 10, wherein the strength distribution is formed based on a distribution in at least one of a temperature of a hardening and a time of a tempering for each of the first and second joining sections.

13. (Withdrawn) A producing method as claimed in Claim 12, wherein the hardening is a high-frequency hardening using an induction heating coil, the hardening being carried out by disposing the induction heating coil along each of the first and second joining sections and by setting a distance between the induction heating coil and each of the first and second joining sections in a manner to form the distribution in the hardening temperature.

14. (Withdrawn) A producing method as claimed in Claim 9, wherein the strength distribution is formed based on a strain introduced into each of the first and second joining sections by a cold forging.

15. (Withdrawn) A method as claimed in Claim 14, wherein the strain gradually and continuously changes with a change in the cross sectional area of each of the first and second joining sections.

16. (Withdrawn) A producing method as claimed in Claim 14, wherein the strain is based on squashing a rib portion of each of the first and second joining sections.

17. (Withdrawn) A producing method as claimed in Claim 14, wherein the strain is adjusted in accordance with a dispersion in thickness of a roughly made connecting rod as a material of the connecting rod.

18. (Withdrawn) A producing method as claimed in Claim 14, wherein each of the first and second joining sections is subjected to an aging after the cold forging.

19. (Previously Presented) A high-strength connecting rod comprising:

a connecting beam section serving as a main body of the connecting rod, the connecting beam section having a smallest cross sectional area portion which is the smallest in cross sectional area throughout the connecting rod;

a big end located at a first end side of the connecting beam section;

a small end located at a second end side of the connecting beam section, the second end side being axially opposite to the first end side;

a first joining section located between the connecting beam section and the big end to connect the connecting beam section and the big end; and

a second joining section located between the connecting beam section and the small end to connect the connecting beam section and the small end;

wherein each of the first and second joining sections gradually and continuously decreases in cross sectional area toward the connecting beam section;

wherein a lowest fatigue strength portion which is the lowest in fatigue strength exists in at least one of the big and small ends, and a variable fatigue strength portion

which varies in fatigue strength exists in each of the first and second joining sections and in the connecting beam section;

wherein a product of the cross sectional area and the fatigue strength at a cross section of each of the joining and connecting beam section is equal to or greater than a product of the cross sectional area and the fatigue strength in the smallest cross sectional area portion in the connecting beam section.

20. (Original) A high-strength connecting rod comprising:

a connecting beam section serving as a main body of the connecting rod, the connecting beam section having a portion which is the smallest in cross sectional area throughout the connecting rod;

a big end located at a first end side of the connecting beam section;

a small end located at a second end side of the connecting beam section, the second end side being axially opposite to the first end side;

a first joining section located between the connecting beam section and the big end to connect the connecting beam section and the big end; and

a second joining section located between the connecting beam section and the small end to connect the connecting beam section and the small end;

wherein each of the first and second joining sections gradually and continuously decreases in cross sectional area toward the connecting beam section;

wherein a cross section of each of the connecting beam section and each of the first and second joining sections includes at least one of martensitic structure and ferritic-pearlitic structure and satisfies the following expression:

$$S/D \geq 1/\{(1-\beta)Ms/100+\beta\} \ldots \text{Eq. (1)}$$

where S is a cross sectional area of any portion of each of the connecting beam section and each of the first and second joining sections; D is a cross sectional area of the smallest cross sectional area portion of the connecting beam section; β is a fatigue strength of an unhardened structure / a fatigue structure of a tempered martensitic structure; Ms is a proportion of area of the tempered martensitic structure in the portion whose sectional area is S;

wherein a whole cross section of the smallest cross sectional area portion is formed of the tempered martensitic structure.

21. (Original) A high-strength connecting rod as claimed in Claim 19, wherein the high strength connecting rod is formed of a steel including, on mass basis, 0.20 to 0.43% of C, 0.05 to 2.0% of Si, 0.30 to 1.40% of Mn, less than 0.07% of P, 2.5% or less of Cr, 0.05% or less of Al and 0.005 to 0.03% of N, and at least one selected from the group consisting of 0.03 to 0.5% of V, 0.005 to 0.5% of Nb and 0.005 to 0.5% of Ti, the balance being Fe and impurities.

22. (Original) A high-strength connecting rod as claimed in Claim 19, wherein the high-strength connecting rod is formed of a steel including, on mass basis, 0.20 to 0.43% of C, 0.05 to 2.0% of Si, 0.30 to 1.40% of Mn, 0.07 to 0.15% of P, 2.5% or less of Cr, 0.05% or less of Al, 0.005 to 0.03% of N, and at least one selected from the group consisting of 0.03 to 0.5% of V, 0.005 to 0.5% of Nb and 0.005 to 0.5% of Ti, the balance being Fe and impurities.

23. (Original) A high-strength connecting rod as claimed in Claim 21, wherein the steel further includes, on mass basis, at least one selected from the group consisting of 2.0% or less of Ni, 1.0% or less of Mo, and 0.0010 to 0.0030% of B.

24. (Original) A high-strength connecting rod as claimed in Claim 21, wherein the steel further includes, on mass basis, at least one selected from the group consisting of 0.2% or less of S, 0.3% or less of Pb, 0.1% or less of Ca, and 0.3% or less of Bi.

25. (Original) A high-strength connecting rod as claimed in Claim 19, wherein the high-strength connecting rod has been subjected to shot peening.

26. (Withdrawn) A method of producing the high-strength connecting rod of Claim 19, the producing method comprising:

forming a material steel into a shape of the connecting rod;

hardening the material steel having the connecting rod shape by using induction current; and

tempering the hardened material steel at a temperature ranging from 200 to 650 °C.

27. (Withdrawn) A producing method as claimed in Claim 26, wherein the tempering is carried out at a temperature ranging from 350 to 550 °C.

28. (Withdrawn) A producing method as claimed in Claim 26, wherein the tempering is carried out by using induction current.

EVIDENCE APPENDIX

No evidence is hereby submitted.

RELATED PROCEEDINGS APPENDIX

There are no related proceedings.

EXHIBIT APPENDIX I

In rejecting a claim, the examiner must set forth express findings of fact which support the lack of written description conclusion (see MPEP § 2163 for examination guidelines pertaining to the written description requirement). These findings should:

(A) Identify the claim limitation at issue; and

(B) Establish a *prima facie* case by providing reasons why a person skilled in the art at the time the application was filed would not have recognized that the inventor was in possession of the invention as claimed in view of the disclosure of the application as filed. A general allegation of "unpredictability in the art" is not a sufficient reason to support a rejection for lack of adequate written description. A simple statement such as "Appellant has not pointed out where the new (or amended) claim is supported, nor does there appear to be a written description of the claim limitation '____' in the application as filed." may be sufficient where the claim is a new or amended claim, the support for the limitation is not apparent, and Appellant has not pointed out where the limitation is supported.

When appropriate, suggest amendments to the claims which can be supported by the application's written description, being mindful of the prohibition against the addition of new matter in the claims or description. See *Rasmussen*, 650 F.2d at 1214, 211 USPQ at 326.

EXHIBIT APPENDIX II

Appellants hereby submit the document *Review of the Performance of High Strength Steels Used Offshore*, prepared by Cranfield University, 2003, as evincing that bainite influences strength (*see, e.g.,* pages 6, 35, 71, 107 and 110, where indicated), as discussed in the Brief.

Appellants previously provided this document with the Response of January, 2006.

Relying on 37 CFR §1.116(e), Appellants submit that this document is contemporary evidence that rebuts the position taken in the Office Action regarding an inherent relation between the presence of martensite in a steel and a strength distribution based on a proportion of martensite. That is, this document is necessary to rebut the arguments ***first*** presented in the Final Office Action of September, 2005, that "standard text books of material science, such as . . . *Mechanical Design and Systems Handbook*, attached clearly teach that the strength distribution is based on a proportion (%) of martiensite" and that therefore "claim 2 is anticipated by JP '317 even though JP '317 does not ***explicitly*** so describe." (September Office Action, paragraph spanning pages 11-12, emphasis added.) Thus, **before** the Final Office Action, **no** allegation was made that the strength distribution was inherently present in JP '317, and *Mechanical Design and Systems Handbook* had **not** yet been relied on by the PTO. Thus, the document *Review of the Performance of High Strength Steels Used Offshore* is necessary to rebut both an assertion and a reference used as a basis proffered by the PTO ***for the first time*** in a Final Office Action to reject claims. Accordingly, sufficient reason exists as to why this document was not earlier presented.

Review of the performance of high strength steels used offshore

Prepared by **Cranfield University**
for the Health and Safety Executive 2003

RESEARCH REPORT 105

Review of the performance of high strength steels used offshore

Professor J Billingham, Professor J V Sharp, Dr J Spurrier and Dr P J Kilgallon
School of Industrial and Manufacturing Science
Cranfield University
Cranfield
Bedfordshire
MK43 0AL

High strength steels (yield strength >500MPa to typically 700MPa) are increasingly being used in offshore structural applications including production jack-ups with demanding requirements. They offer a number of advantages over conventional steels, particularly where weight is important. This review considers the types of steel used offshore, their mechanical properties, their weldability and their suitability for safe usage offshore in terms of fracture, fatigue, static strength, cathodic protection and hydrogen embrittlement performance. In addition, this review addresses the performance of high strength steels at high temperatures and at high strain rates. It outlines the difficulties in working with the very limited published codes and standards and discusses performance in the field. Current design restrictions such as limits on yield ratios, susceptibility to hydrogen cracking including the influence of SRBs, and the management of the behaviour of such steels in seawater under cathodic protection conditions are discussed. Recommendations are made to encourage the wider use of high strength steels in the future and areas where further study is required are identified.

This report and the work it describes were funded by the Health and Safety Executive (HSE). Its contents, including any opinions and/or conclusions expressed, are those of the authors alone and do not necessarily reflect HSE policy.

HSE BOOKS

First published 2003

ISBN 0 7176 2205 3

CONTENTS

NOMENCLATURE

a	Crack length parameter
B	Thickness of fracture specimen
C_E or CE	Carbon equivalent
CP	Cathodic protection
CR	Controlled rolled
CTOD	Crack tip opening displacement
C_v	Charpy impact energy
□	Crack tip opening displacement value
da/dN	Crack growth rate
□K	Stress intensity factor range
□	Young's modulus
□□	Embrittlement index
FCAW	Flux cored arc welding
FCGR	Fatigue crack growth rate
FMD	Flooded member detection
HAC	Hydrogen assisted cracking
HAZ	Heat affected zone
HE	Hydrogen embrittlement
HIC	Hydrogen induced cracking
HSLA	High strength low alloy
HSS	High strength steels
HV	Vickers hardness
ICCP	Impressed current cathodic protection
ISO	International Standards Organisation
J	Joules
K_{app}	Applied stress intensity factor
K_c	Apparent toughness
K_{1A}	Arrest toughness
K_{1c}	Material toughness
K_{1D}	Stress intensity factor to keep crack in motion
K_{ISCC}	Fracture toughness under conditions of stress corrosion cracking
K_{th}	Threshold value of K
LBZ	Local brittle zones
MAC	Martensite-austenite content
MMA	Manual metal arc
MPa	Mega Pascals
Q&T	Quench and tempered
□$_y$	Yield stress
R curve	Resistance curve (energy per unit area of crack extension)
R_e	Specified min. yield strength (in fracture toughness equations)
SACP	Sacrifical anode cathodic protection
SAW	Submerged arc welding
SMYS	Specified minimum yield strength
S-N	Stress versus no. of cycles in fatigue
SRB	Sulphate reducing bacteria

SSCC	Sulphide stress corrosion cracking
SSRT	Slow strain rate testing
TLP	Tension leg platform
TMCP	Thermomechanically controlled processing
TMCR	Thermomechanically controlled rolling
TT	Ductile to brittle transition temperature
UKCS	UK Continental shelf
UTS	Ultimate tensile strength
□	Poisson's ratio
YR	Yield ratio (□$_y$/UTS)
YS	Yield strength

SUMMARY

High strength steels (yield strength >500MPa to typically 700MPa) are increasingly being used in offshore structural applications including production jack-ups with demanding requirements. They offer a number of advantages over conventional steels, particularly where weight is important. This review considers the types of steel used offshore, their mechanical properties, their weldability and their suitability for safe usage offshore in terms of fracture, fatigue, static strength, cathodic protection and hydrogen embrittlement performance. In addition, this review addresses the performance of high strength steels at high temperatures and at high strain rates. It outlines the difficulties in working with the very limited published codes and standards and discusses performance in the field. Current design restrictions such as limits on yield ratios, susceptibility to hydrogen cracking including the influence of SRBs, and the management of the behaviour of such steels in seawater under cathodic protection conditions are discussed. Recommendations are made to encourage the wider use of high strength steels in the future and areas where further study is required are identified.

1. INTRODUCTION

Fixed offshore structures are conventionally constructed from medium grade structural steels, with yield strengths typically in the range of 350MPa. These steels are well documented and covered by existing codes and standards. However, in recent years there has been an increasing interest in the use of higher strength steels for these installations, recognising the benefits from an increase in the strength to weight ratio and the associated savings in the cost of materials. As a result, significant parts of several platforms (jacket and topsides) have been constructed from 400 – 450MPa steel and installed in the North Sea. However, to date, fatigue sensitive components (e.g. tubular joints) have generally been fabricated from medium strength steel because of the better knowledge on these steels regarding fatigue performance and the lack of increased performance of high strength steels in this area.

The principal application of very high strength steels offshore has been in the fabrication of jack-ups. Steels with nominal yield strengths in the range 500 – 800MPa are normally used in fabrication of legs, rack and pinions and spud cans. These units, used primarily for drilling, have many years of satisfactory experience in use, operating in a variety of water depths, but are normally brought into dry dock for inspection at 5 year intervals, where any damage or cracking can be found and repaired. In recent years there has been increasing interest in the use of jack-ups for production, where periodic dry dock inspection is not possible. Two jack-ups for production, utilising high strength steels, are now installed (Harding in 1996, Siri in 1998) and a third (Elgin-Franklin) is due to be installed shortly on the UKCS. High strength steels have also been used in tethering attachments for floating structures in TLPs (tension leg platforms) and for mooring lines with semi-submersible module offshore drilling units (MODUs).

A considerable amount of research has been undertaken on high strength steels in recent years providing new data to support offshore applications. However, overall there is limited information of the long-term use of high strength steels in seawater, particularly under the severe environment conditions to which structures on the UKCS are subjected. Particular concerns with the use of higher strength steels are the greater susceptibility to hydrogen cracking which can be enhanced when SRBs are present, their fatigue and fracture performance, and, for offshore applications, their performance at higher temperatures as a result of fire.

Most codes and standards relate to medium strength steels and in most cases the use of design formulae is limited to steels with yield strengths <500MPa which is a serious disadvantage for the use of higher strength steels. Despite the increasing amount of data available from research and testing, very little of this has yet found itself into codes and standards. The application of current codes and standards to high strength steels will be reviewed in this report.

Several reviews of high strength steels offshore have been produced and published between 1995 and 1999 [1.01-1.08]. The plan adopted for this review is to use the information in these as a basis, but to include new data, applications and codes and standards since these were produced. In particular, where possible, the performance of steels with a yield strength of 450MPa, where there is some published data, will be used as a baseline for assessing the performance of higher strength steels. In some cases the performance of medium grade steels (YS ~350MPa) will need to be used as a benchmark since limited data are available for higher strength steels. This review covers a wider scope than previously published reviews of high strength steels in that it includes fire and impact resistance as well as field performance and inspection and maintenance aspects. High strength steels up to 450MPa (X70 grade) have been used for many years in offshore pipelines. However, pipelines, in general, have different design criteria and requirements from offshore structures. Pipelines were therefore not included in the current review except where general principles of steel metallurgical development or the like were a common feature to both types of application.

REFERENCES

1.01 *Billingham J, Healy J, and Spurrier J,* 'Current and potential use of high strength steels in offshore structures', Publication 95/102 MTD, Sept. 1995, 51 pages, ISBN 1-870553-24-1

1.02 *Billingham J, Healy J, and Bolt H,* 'High strength steels – the significance of yield ratio and work hardening for structural performance', Marine Research Review 9, 36 pages, published MTD 1997, ISBN 1-870553-27-6

1.03 *Billingham J, Sharp J V, Spurrier J, and Stacey A,* 'The use of high strength structural steels in offshore engineering', Intnl.Symposium on Safety in Application of High Strength Steel, Trondheim, July 1997

1.04 *Sharp J V, Billingham J, and Stacey A,* 'Performance of high strength steels used in jack-ups in seawater' Marine Structures, Vol.12, 1999, 349-369, ISSN 09 51-8339, Elsevier 1999

1.05 *Healy J, and Billingham J,* 'A review of the corrosion fatigue behaviour of structural steels in the strength range 350-900MPa, and associated high strength weldments', Offshore Technology Report, OTH 532, Publ. Health & Safety Executive, 1997, ISBN 0-7176 2409-9

1.06 *Robinson M J, and Kilgallon P J,* 'Review of the effects of sulphate reducing bacteria in the marine environment on the corrosion fatigue and hydrogen embrittlement of high strength steels', Offshore Technology Report, OTH 555 98, HSE Books (1999), Sudbury, Suffolk

1.07 *Robinson M J, and Kilgallon P J,* 'Review of the effects of microstructure on the hydrogen embrittlement of high strength offshore steels', Offshore Technology Report (in press – no number yet issued) HSE Books (1999), Sudbury, Suffolk

1.08 *Stacey A, Sharp J V and King R N,* 'High Strength Steels used in Offshore Installations', Proceedings 15th International Conference Offshore Mechanics and Arctic Engineering, Florence, Italy, June 1996, Vol III

2. THE USE OF HIGH STRENGTH STEELS OFFSHORE

Traditionally, offshore structures have been fabricated with moderate strength steels with yield strengths up to 350MPa [2.01], mainly produced by the normalising route. However, there has been a significant growth over the past twenty years in the use of high strength steels in the offshore engineering industry, primarily driven by a desire to save weight and cost [2.02]. Table 2.1 shows the main application areas involved which vary with the strength of steel used. Such steels are also generally produced by alternative processing routes such as thermomechanical controlled processing (TMCP), and quenching and tempering (Q & T).

The principal advantage of using these structural materials is their increased strength to weight ratio and the attendant savings in materials costs and construction schedules [2.03] due to reduced amounts of welding. The most important increases have occurred in the topside areas of jacket structures where the weight saving has not only produced overall savings in materials used but has allowed crane barge installation of more complete topside processing and accommodation units [2.04] with significant related cost savings. A survey undertaken in 1995 [2.05] indicated that the proportion of high strength steel (defined as >350MPa yield strength) used in offshore structures increased from less than 10% to over 40% over little less than a decade.

More recent applications, especially with smaller, lighter structures, involve the use of such steels in the jacket members themselves although there are still usually restrictions to their use in nodal areas because of concerns related to fatigue performance. It is likely that the use of such materials will continue to increase as the steels become more widely available and construction yards get more experienced in fabrication procedures. To date, most steels have been restricted to 450 grades but research and development programmes [2.06] have indicated that steel grades up to 550MPa can be produced which are readily weldable and possess excellent fracture toughness. Such steels will increasingly be utilised as they become more widely available.

Higher strength steels (>550MPa and often up to 700MPa) are usually produced by the quenching and tempering route and have traditionally been used offshore in mobile jack-up drilling rigs which do not stay permanently on station and are periodically dry docked, allowing inspection and repair programmes to be implemented [2.07]. The principal application of high strength steels in jack-ups is in the fabrication of the legs because of the requirement to minimise weight during the transportation stage. In general, each lattice leg is composed of three or four longitudinal chord members which may contain a rack plate for elevating the hull and a series of horizontal and diagonal tubular braces which connect the chords to form a truss. Supplementary braces (span breakers) are frequently used between main brace mid-points to increase the buckling resistance. The rack plate is very thick, varying typically between 150 and 250mm. The chord shell cans are usually fabricated from plate with a wall thickness between 35 and 80mm and with a diameter in the range 800 to 1200mm. Weldability and good toughness and ductility are important material considerations in this application and the steel maker provides this by careful control of alloy composition and by processing [2.01; 2.04].

Steels of similar strength levels have only comparatively recently been used in production jack-ups permanently on station in North Sea projects in the Harding and Siri fields, and in the Elgin jacket which was installed in 2001. In such installations, fatigue, corrosion fatigue and hydrogen embrittlement become major design considerations and the steels used have to be carefully reassessed. The French TPG 500 design is a good example of this type of structure. It can be built onshore as one complete unit and floated out to site. Once on station the legs can be lowered to the seabed and the deck jacked up for operation, thus reducing the need for heavy lift operations during installation and producing significant cost savings. A second benefit in this design is that it is a reusable production facility, since it can be refloated, removed from one site to another, and commence operations in the new field. To provide the required fatigue life the legs of the structure have incorporated forged

nodes (fabricated by Creusot-Loire), thus reducing the stress concentrations normally seen in welded nodes. The Harding platform is in 110m water depth and the lower part of the structure comprises a concrete base to remove any potential problems of hydrogen embrittlement related to sulphate reducing bacteria in the mud zone. A jack-up was also installed as a permanent installation in the Danish sector (Siri field, 60m depth) in 1998. The majority of the leg sections on the Siri platform are made from thick section (65-110mm) 690 grade high strength steel. The Elgin structure uses a range of high strength steels including 500MPa steel in structural members, chords and bracings and 700MPa steel in the racks. It uses lower strength steels (350MPa) in the lower part of the structure which are piled into the seabed to avoid potential hydrogen embrittlement problems.

Other applications for high strength steels are found in mooring attachments for floating structures such as tension leg platforms (TLPs). These structures are fixed by vertical tension legs to piled foundation templates on the seabed. One of the first such designs for the Hutton field in the UK sector used 16 tension legs (4 at each corner). Each leg is a thick walled steel tubular, manufactured from a low alloy steel (3.5%Ni, Cr, Mo, V) with a minimum yield strength of 795MPa [2.08]. The individual components of each leg are forged and connected by screwed couplings. The steel composition was selected to provide the highest possible strength, commensurate with adequate fracture toughness. Resistance to both stress corrosion cracking and corrosion fatigue were also important. The choice of screwed connectors was based on the fact that there were insufficient data available on the corrosion fatigue performance of welded tubulars to guarantee safe performance under the envisaged design life of the structures. A large test programme was undertaken to justify the choice of material. The platform has now been in operation for almost 20 years without any significant problems with the tethers.

Since then TLP type platforms have been installed in several other fields, both in Norway and in the Gulf of Mexico. In the Heidron field, for example, a welded TMCP (thermo-mechanically processed) microalloyed X70 pipeline steel was used for the tethers. The structure is in 270m water depth and comprised tubular tension leg elements that were 1m diameter and 38mm thick. The steel had a yield strength of 500MPa and impact toughness of 60J at -40°C. Other structures have been used in much deeper waters, mainly in the US (Auger TLP in 872m, installed in 1994; Mars TLP in 896m of water in 1996), where conventional fixed platforms are uneconomic. These projects involved 76mm thick 415MPa components of weldable TMCP steel. Plans are in hand for even deeper water TLP units, but it is now recognised that the availability of suitable high strength materials for the tethers is limiting further development.

Other floating structures such as semi-submersibles, used welded higher strength steel anchor chains or wire ropes as their mooring attachments. Such steel chain and wire rope components are considered outside the scope of the present review.

Other, more specialised, usage areas include flanges and repair clamps where threaded fasteners provide the main load transfer mechanism. Such bolts and threaded fasteners are discussed in more detail in Appendix I.

Table 2.1
High strength steels used offshore

Strength MPa (grade)	Process Route	Application Area
350 (X52)	Normalised TMCP	Structures Structures & Pipelines
450 (X65)	Q & T TMCP	Structures Pipelines
550 (X80)	Q & T TMCP	Structures & Moorings Pipelines
650	Q & T	Jack-ups & Moorings
750	Q & T	Jack-ups & Moorings
850	Q & T	Jack-ups & Moorings

REFERENCES

2.01 *Billingham J,* 'Steel – a versatile advanced material in marine environments', Ironmaking and Steelmaking 1994, Vol.21, No.6, p422

2.02 *Billingham J, Healy J, and Spurrier J,* 'Current and potential use of high strength steels in offshore structures', MTD Publication 95/102 (1995), ISBN 1-870-533-24-1

2.03 *Rodgers K J, and Lockhead J C,* 'The welding of grade 450 offshore structural steels', Proc Conf. on Welding and Weld Properties in the Offshore Industry, London, April 1992

2.04 *Webster S E,* 'Structural materials for offshore structures – past, present and future', Proc.Conf. on Safe Design and Fabrication of Offshore Structures, IBC, London, Sept. 1993

2.05 *Healy J, and Billingham J,* 'High strength steels – a viable option for offshore designs', Euroforum Conference – Latest Innovations in Offshore Platform Design and Construction, London 1996

2.06 'The influence of welding on materials performance of high strength steels offshore', Managed Programmes of University Research, Marine Technology Centre, Cranfield University, 1985-1994

2.07 *Sharp J V, Billingham J, and Stacey A,* 'Performance of high strength steels used in jack-ups" Journal of marine Structures, 12 (1999), 349

2.08 *Salama M N, and Tetlow J H,* Proc. of Offshore Technology Conference, Houston, Texas, 1983, Paper 449.

3. MECHANICAL PROPERTIES OF HIGH STRENGTH STEELS

3.1 STEEL TYPE AND PROCESS ROUTE

In general, the strength of a steel is controlled by its microstructure which varies according to its chemical composition, its thermal history and the deformation processes it undergoes during its production schedule. In addition, structural steel for offshore applications must be readily weldable since this is the traditional fabrication route for offshore structures. Structural steels for offshore must therefore be available in moderate to thick sections (30 – 100mm) and must exhibit good toughness to avoid the possibility of brittle failure, in addition to showing good weldability and high strength. Such overall requirements are often difficult to achieve because an increase in one of these properties often leads to a decrease in the others.

Table 3.1 below shows the strength ranges and process routes for high strength steels used in a variety of offshore engineering applications.

Most conventional structures use only moderate strength steel produced by the normalised or thermomechanically processed routes (TMCP) but at higher strength levels there are processing thickness restrictions to TMCP steels and normalising cannot produce the strength levels required in the necessary section thicknesses. Quenching and tempering is therefore the standard production route for very high strength structural steel. The limitations that apply to the different process routes in respect of strength or thickness ranges are shown in Table 3.2.

3.2 METALLURGICAL AND COMPOSITIONAL CONSIDERATIONS

The offshore pipeline industry, for many years, has used high strength steels and today commonly uses X70 steel grade (/450MPa) with excellent toughness and weldability properties [3.01]. Significant benefits in such developments have come from understanding the complex chemistries developed for the steel plus the use of extensive thermomechanical processing, primarily to produce fine grained microstructures, including controlled rolling, thermomechanical controlled processing and accelerated cooling. Many of the principles involved in such developments, particularly the complex interactions between strength, toughness and weldability as influenced by steel chemistry, heat treatments and thermal processing [3.02] have been carried over into higher strength steel development.

Many of these well understood metallurgical principles can be utilised to satisfy the overall mechanical property requirements for high strength structural steels, namely:

- reduced carbon content to improve weldability and toughness;
- decreased grain size (ferrite and/or bainite) to give increased strength and increased toughness. This is usually achieved by microalloying with Nb, V or Al and by some form of thermomechanical processing;
- decreased impurity content (S, P, O) to increase toughness in particular and through thickness homogeneity, i.e. the use of clean steel technology;
- increased alloying with Ni, Cr, Mo and Cu to give solid solution and transformation strengthening, especially at the higher strength levels.

Relatively small changes in composition and/or variations in processing route can significantly affect the resulting mechanical properties as shown in Table 3.3. This table shows 'old' and 'new' versions of steels within 3 standard steel grades, i.e. 355, 450 and 690MPa yield strength. Although all of the steels within a particular grade satisfy the grade requirements (primarily with respect to specified minimum yield strength) it can be seen that the 'newer' versions show much improved overall properties by combining the required yield strength with improved toughness (improved Charpy impact performance {Cv}), and improved weldability (lower carbon equivalent values [C_E[1]}). This

[1] Carbon equivalent CE is defined as CE=C+Mn/6+(Cr+Mo+V)/5+(Ni+Cu)/15 – See page 30 for more details.

has been achieved primarily by controlling the microstructure through changes in chemistry and thermal processing.

Many engineers and designers do not appreciate that the mechanical properties of a particular steel can vary significantly within a specified steel grade (i.e. steel with a specified minimum yield strength). Figure 3.1 shows the variation in mechanical properties for three offshore steel grades with minimum yield strengths of 355, 420 and 450MPa [3.03]. For the 450MPa steel, for example, it can be seen that the yield strength can vary by 100MPa from 440 to 540MPa (+20% on design yield value), with a mean value at approximately 500MPa. Such variations are produced by variations in steel composition and processing which affect all the mechanical properties as shown in Figure 3.2 for a 450MPa steel. Such variations can have serious implications for the degree of weld overmatching or undermatching that occurs in the final structure. The range of properties achievable within a particular grade can also vary significantly with process route as shown in Figure 3.3 which illustrates the much wider variation obtained from the TMCP route than from either the normalised route at lower strength levels or the quenched and tempered route at higher strength levels [3.04]. Variations can also occur with plate thickness and with steel manufacturer [3.05]. It is important that this potential variation in yield strength is recognised at the design stage.

Typical compositions and properties for high strength steels produced by the normalised, thermomechanically controlled processed and quenching and tempered routes are shown in Tables 3.4, 3.5 and 3.6 respectively. Other typical composition and mechanical properties are given in the draft DNV Offshore Standard OS-B101 Metallic Materials, (May 2000). In general for such steels the strength increases as the hardenability and the composition related carbon equivalent values increase (see Figure 3.4). for each process route, but the particular process route selected generally has a more significant influence on yield strength. By making use of strength improvements associated with the processing route, it is possible to produce steels of the same strength level at leaner chemistries and hence lower carbon equivalent values, which show improved weldability. By using careful control of composition and processing, steels can therefore usually be produced with excellent combinations of strength and toughness combined with excellent weldability. In general, as the strength increases the weldability in particular decreases and more control over the welding procedures such as increased levels of preheating are usually required. Moreover, in general, the toughness of very high strength steels (690MPa) is inferior to that of steels with low (350MPa) or intermediate (450-550MPa) levels of strength.

3.3 YIELD RATIO CONSIDERATIONS

The stress strain behaviour of high strength steels differs somewhat from that of lower strength steels in that they generally show reduced capacity for strain hardening after yielding and reduced elongation as shown in Figure 3.5. This is because the steel strengthening mechanisms used in high strength steel development have been selected specifically to increase the yield strength and have much less influence in subsequent strain hardening behaviour. One measure to illustrate this different behaviour is yield ratio (YR) which is defined as the ratio of yield strength (σ_y) to ultimate tensile strength (UTS), and which generally increases as the strength of the steel increases as shown in Figure 3.6 for a range of offshore steel grades [3.06]. YR is not, however, a unique measure of how the steel behaves because steels with very different stress strain curves can have the same value of YR [3.06]. There are restrictions in structural design codes to reflect this changed behaviour such that YR for material to be used for structural members is not allowed to have a value greater than 0.85 in design equations to ensure that there is adequate ductility in the member to develop plastic failure behaviour as a defence against brittle fracture. Design aspects related to YR are given in section 13 of this report. Examination of a different database [3.07] shown in Figure 3.7 shows that generally steels with yield strengths up to 500MPa can satisfy this general requirements but that very high strength steels do not.

The yield ratio is not directly related to the capability of a given steel to withstand plastic strain after yield and before fracture. In older high strength steels, elongation generally decreases as yield ratio

increases, but modern clean steels with low carbon content and low levels of impurity have significant elongation even at the highest strength (690 grade) and yield value ratio (0.95), giving more confidence as to their deformation capability [3.05]. An alternative measure is the general elongation which is usually substantial in modern steel with high yield ratios.

Current design equations are based on test data from medium strength steels, where some degree of strain hardening is present. Lack of strain hardening can lead to premature cracking, which could have significant implications for tubular joints in service. As a result, design codes have placed a limit on the yield ratio (typically 0.7). Examination of Figure 3.7 emphasises the range of values that can occur within particular strength grades, largely due to the different methods of production, differences in steel chemistry, and differences in section thickness that occur (see section 3.2). Indeed, from this diagram it can be seen that 350MPa steels generally show a yield ratio ranging from 0.6 to 0.8, that 450MPa steels have values ranging from 0.7 to 0.87, whereas 690MPa steels have values ranging from 0.9 to 0.95. Elongation generally decreases in line with increasing yield ratio; therefore for 350 – 450 steels, elongations are generally of the order of 20 – 35%, whereas for 690 steels, values of 14 – 18% are more typical [3.06].

Examination of Figure 3.7 shows that many steels, even at strength levels up to 400MPa have yield ratios above 0.7, which could include many steels purchased at grade 355 level. It is therefore possible that some earlier structures might have nodal joints which do not satisfy the current design codes although they have performed perfectly satisfactorily in service.

There has for some time been a feeling that the code restrictions for nodal connections are rather conservative in respect of high strength steels because intuitively it would be expected that joint load capacity would increase in line with yield strength, whereas the code restrictions impose severe limitations. For example, on increasing the yield strength from 355 to 532MPa (a 50% increase) the designer is only allowed to increase the allowable design stress by 23% when the yield ratio is 0.85 (YR = 0.85, design stress = 0.7UTS = 0.7 x 626 = 438MPa). Other examples of the restrictions imposed by the code are given in Appendix 3. Initial finite element studies on X joint deformation behaviour by BOMEL and Cranfield [3.06] indicated that joints with high YR had significantly higher joint capacities than joints with low ratios. For example, a joint with a YR of 0.93 (490MPa σ_y, 525MPa UTS) showed a 28% increase in joint capacity compared to a lower strength steel σ_y = 350MPa with a YR of 0.66 and the same value of UTS (525MPa UTS). Existing structural codes would have restricted the capacity of both these joints to the same value (YR = 0.66 $\sigma_D = \sigma_y = 350$, YR 0.93, $\sigma_D = 2/3 \times \text{UTS} = 2/3 \times 525 = 350$). Despite the above enhanced in-joint capacity, the capacity does not increase linearly with yield stress as indicated by the design equations (static strength for DT joints):

$$P = F_y \ \text{MIN}\left(1, \frac{0.7}{\text{YR}}\right) T^2 \ (2.5 + 14\beta) \ Q_\beta$$

where P = static design strength, F_y = yield strength, YR = yield ratio, T = wall thickness, β = diameter ratio d/D, and Q_β is a geometric factor, defined as $Q_\beta = 0.3/\beta(1-0.833\beta)$ for $\beta > 0.6$ and Q_β = 1 for $\beta < 0.6$). In the case quoted a direct dependence on yield strength would have indicated a 40% increase in capacity rather than the measured increase of 28%. However, on the basis of this work, the authors concluded that the current restriction of 0.7 could possibly be relaxed to 0.8. Other workers, notably Healy and Zettlemoyer [3.08], and Wilmshurst and Lee [3.09] have also indicated that they thought the current restriction was too severe and should be relaxed.

An analysis of the limited static strength data for tubular joints manufactured from high strength steels (>600MPa) [3.10] showed that the recommended restriction of 0.7UTS was justified. However, the analyses were carried out using the HSE Guidance Notes equations where ultimate strength is the failure criterion. The point was also made that the data were very limited and that the range of joint

types tested was limited (1 T, 8 K/YT, 2 DT). In addition, the geometry range was limited (lowest gamma was 14.8, highest beta 0.43). The range of yield ratios was 0.88 to 0.94.

A later study for the European Commission [3.11] which involved some relatively large scale experimental tests by BOMEL, TNO and Delft Universities largely confirmed these earlier experiments. This programme included a significant finite element study, a comprehensive re-examination of the test database used in setting up the structural code formulations and an expanded experimental test programme involving 2 series of tests (compression and tension) on DT joints made from 350, 450 and 690MPa steels. The finite element analyses successfully reflected the experimental DT test joint data. They also isolated and quantified accurately the effect of geometric imperfections in the test joint. The earlier indications that the design stress for tubular joints should be raised from the YR = 0.7 value in the guidance were confirmed. The authors concluded that the conservatism utilised in high strength steel tubular steel design could be reduced by changing the YR = 0.7 limit in the design equation to 0.8 for both compression and tension joints. This would then enable more widespread use to be made of high strength steel in tubular joint design. The authors also recommended that more tests on tubular joints, especially at the higher grades of steel and especially in tension, were required.

A recently published paper [3.12] from Thyssen Krupp Stahl AG analysed transition temperatures from Charpy V-tests and the fracture mechanics transition temperature for 690MPa steels against yield ratio and found no correlation, but it was recognised that these results were from small scale tests. Another analysis of the maximum net stress versus test temperature of the wide plate tests for 690MPa steels, with yield ratios ranging from 0.87 to 0.93 showed that the highest loads were in the steels with the highest yield ratio. The authors concluded that yield ratio is not a good measure of component safety, and that other factors should be taken into account. This paper [3.12] also listed limitations on yield ratio in various design codes and materials standards (both onshore and offshore) ranging from 0.7 to 0.93 for various components.

Since 1996, several Panels have been meeting to draft a new ISO standard for offshore structures. This includes a Panel drafting a section on the static strength of tubular joints. The Panel has re-examined the test data on joint strength and developed some improved design equations. However, because of the lack of data on higher strength steel joints, the Panel concluded that these equations should be limited to steels with yield strengths less than 500MPa (see section 4). However, even for these steels, it was considered necessary to impose a limit of the yield ratio (since some lower strength steels have yield ratios greater than 0.7). The Panel concluded that on the basis of test results for lower strength steels, the limiting yield ratio should be 0.8.

For higher strength steels (yield strength greater than 500MPa) the Panel concluded that use of a limiting value of yield ratio of 0.8 may be adequate to permit the ultimate compression capacity equations to be used for joints with strengths in the range 500-800MPa provided adequate ductility can be demonstrated in the HAZ and parent material. It is unclear how this demonstration of adequate ductility can be provided either in terms of mechanical property data of the steels concerned or in identified test procedures.

Later examination of some of the available static strength data [3.13] has concluded that although the factor for compression loading could be relaxed to 0.8, the factor for tension loading should, indeed, be lowered to 0.5 based on the design capacity being related to first cracking rather than to ultimate strength as in, for example, the API RP2A code. Failure modes in the compression tests involved an indentation of the chord of about 30% of the diameter. Cracks appeared in the tension specimens at loads of around 50% of the maximum load reached in the tests. However, it should be recognised that these recommendations are based on very limited test data.

Overall, the design of high strength welded joints for static strength is unclear based on the very limited existing data. When failure is defined as the onset of cracking (under tension loading, e.g. as in API RP2A) it would appear that the existing design equations are unconservative for high strength

steel joints, even with a yield ratio of 0.7. For compression loading a limiting yield ratio of 0.8 would appear appropriate, provided there is adequate ductility present in both the HAZ and parent plate. Further data are required to resolve these uncertainties.

An alternative approach to the concerns regarding the influence of high yield ratio and deformation capacity of high strength steel tubular joints is to redesign the steel and its production route to develop high strength steels with lower yield ratios. Japanese studies [3.14] aimed at developing steels for earthquake resistant structures have utilised accelerated cooling and intercritical quenching procedures to produce a microstructure of ferrite dispersed in a bainitic matrix which can produce high strength steel 500MPa in thick sections with a yield ratio of 0.7 and CE values of 0.4 which would indicate reasonable weldability. 700MPa steels with YR of only 0.83 have also been developed but these are less weldable (CE 0.52) [3.15]. The weldability of such steel is inferior to modern HSLA steels used offshore but could almost certainly be improved with further development.

Castings can offer advantages over welded structural fabrications because the joint intersections can be easily contoured to reduce stress concentration effects, at say nodal joints for example, with a corresponding increase in fatigue life [3.16]. In conventional welded nodal joints the fatigue life is decreased because of the microcracking that exists in the weld toe which is eliminated in cast joints with a corresponding significant increase in fatigue life as shown in Figure 3.8. High strength steel castings are available which have attractive combinations of strength and toughness properties [3.17; 3.18]. Steels with yield strengths up to 690MPa are available. They cannot derive their strength from processing so usually have addition of nickel and chromium to suppress transformation temperatures and produce low carbon martensitic or bainitic structures. The excellent through thickness properties of castings have also opened up new markets in lifting attachments, spreader bars, and pad eyes [3.16].

REFERENCES

3.01 'Steels for Linepipe and Pipeline fittings', Proc.Intnl.Conf. London 1981, publ Metals Society

3.02 *Billingham J,* 'Steel – A Versatile Advanced materials in marine Environments', Ironmaking and Steelmaking, Vol.21, No.6, 452, 1994

3.03 *Billingham J, Healy J, and Spurrier J,* 'Current and Potential Use of High Strength Steels in Offshore Structures, MTD publication 95/102, published 1996, ISBN 1-870553-24-1

3.04 *Denys R,* Conf Evaluation of Materials in Severe Environments, ISI Japan, Vol.2, 1989.

3.05 *Healy J, and Billingham J,* Metallurgical Considerations of the High Yield to Ultimate Ratio in High Strength Steels for Use in Offshore Engineering', 14th Intnl.Conf. OMAE 1995, Vol.III, 365

3.06 *Billingham J, Healy J, and Bolt H,* 'High Strength Steel – the Significance of Yield Ratio and Work Hardening for Structural Performance', Marine Research Review 9, publ MTD 1997, ISBN 1-870553-27-6

3.07 *Willcock R T S,* 'Yield : Tensile Ratio and Safety of High Strength Steels', HSE Report, Mat R108, 1992

3.08 *Healy B E, and Zettlemoyer N,* 'In plane bending strength of circular tubular joint', Proc. 5th Intnl.Symposium on Tubular Structures, ed. M G Coutie and G Davies, E and F N Span, 1993

3.09 *Wilmhurst S R and Lee M M K,* Non-linear FEM Study of Ultimate Strength of Tubular Multiplaner Double K Joint', Proc 12th OMAE, Glasgow 1993.

3.10 *Lalani et al,* ESCS Report 7210 MC/602 1996

3.11 Static Strength of High Strength Steel Tubular Joints, ECSC 7210 MC/602 1996, reported in *Wicks P J, and Stacey A,* 'Static Strength of High Strength Steel Tubular Joints', Proceedings ETCE/OMAE 2000 Conference, Paper 2081, February 2000, Publ, ASME

3.12 *Kaiser H J, Kern A, Niessen T, and Schriever,* 'Modern High Strength Steels with Minimum Yield Strength up to 690MPa and High Component Safety', Proc. 11th Intnl. Offshore and Polar Engineering Conference, Norway 2001, ISBN 1-880653-55-9.

3.13 Private Communication with N W Nicholls

3.14 *Shikani N, Kurihora M, Tagawa H, Salkui S, and Watanabe I,* 'Development of high strength steel with low yield ratio for large scale steel structures', Proc. of Microalloying 88, Chicago 1988, p481

3.15 *Toyoda M,* 'Strain Hardenability of High Strength Steels and Matching Properties in Welds', 8th Intnl. OMAE Conf. 1989

3.16 *Marston G J,* 'Novel Application of Structural Steel Castings in the Offshore Industry' The Safe Design and Fabrication of Offshore Structures, IBC Conf., London, 1993

3.17 *Richardson R C,* 'Higher Strength Cast Steel for Offshore Structures', World Expro 161

3.18 *Cowling M J,* 'Fatigue Performance of Cast Steel Intersection for Offshore Structures, in Fatigue Crack Growth in Offshore Structures, ed. W D Dover, S Dharmavason, F P Brennan and K J Marsch, EMAS 1995.

Table 3.1 Strength ranges and process routes for high strength steels used in offshore engineering

Type of structure	*Strength levels used (MPa)*	*Process Route*
Jacket structures and topsides	350 – 500	Normalised Q & T TMCP
Pipelines	350 – 550 (X52) (X80)	TMCP
Jack-ups/Moorings	500 – 850	Q & T

Table 3.2 Steel processing routes for production of high strength structural steels

Normalised	Usually <460MPa for 50mm plate
Thermomechanically controlled rolled (TMCR)	Thickness restriction especially at higher strengths – usually less than 550MPa at 40mm
Accelerated cooled (TMCP)	Improved properties compared to TMCR but thickness restriction at higher strengths
Quenched & Tempered (QT)	(a) Alloyed route – no real thickness restriction but expensive and costly to weld (b) Microalloyed route – thickness and strengths required offshore can be produced
Castings	Usually alloyed because of lack of processing capability

Table 3.3
Effect of changes in processing and alloying methodology on mechanical properties of Grade 355, 450 and 690 steel plates

Steel design	Process	Chemical Composition												Yield Strength (MPa)	Cv Impact Toughness	Weldability (CE_{IIW})
		B	C	Mn	Si	Ni	Cr	Mo	Cu	S	P	Al	V			
Grade 355 BS4340 50D	Normalised OLD	-	0.20	1.35	0.30	-	-			0.016	0.015	0.02	-	360	70J @ -40°C	0.43
BS7191 355EMZ	Normalised NEW	-	0.11	1.50	0.40	0.15	0.15			0.005	0.015	0.03	-	380	>200J@-40°C	0.39
BS4360 50D	TMCP	-	0.07	1.49	0.21	0.38	0.02			0.002	0.008	0.02	-	380	200J @ -30°C	0.36
Grade450 Q1N	Q & T OLD	-	0.18	0.4	0.30	3.0	1-1.8			0.015	0.005	0.02	0.02	550	80J @-85°C	0.81
BS7191 450EMZ	Q & T NEW	-	0.11	1.49	0.3	0.52	0.11			0.001	0.010	0.03	-	480	300J @-40°C	0.40
Dillinger 450TMCP	TMCP	-	0.09	1.50	0.3	-	-			0.001	0.007	0.03	0.04	500	300J @-30°C	0.35
Grade 690 Q2N	Q & T OLD	-	0.11	0.42	0.23	3.40	1.48	0.46	0.03	0.001	0.012	0.026	0.08	550 – 690	80J @-84°C	0.81
OX812	Q & T NEW	-	0.11	0.89	0.26	1.18	0.46	0.38	0.15	0.003	0.008	0.07	0.01	690	100J @-80°C	0.52
SE702	Q & T NEW	0.0027	0.125	1.05	0.25	1.4	0.5	0.45	0.20	<0.002	<0.01	-		750	120J @ -40°C	0.59
DSE 690V	Q & T NEW	-	0.15	0.90	0.33	1.28	0.49	0.45	0.2	0.001	0.009	0.073	0.03	700	74J @ -60°C	0.59

Table 3.4

Typical composition and mechanical properties of normalised steels produced in Europe – yield strength range 350 to 490MPa

Thickness *(mm)*	*Typical composition (by weight %)* *C*	*Mn*	*Si*	*S*	*P*	*Nb*	*V*	*Al*	*Cu*	*Ni*	*Cr*	*Mo*	CE_{IIW}	*Typical mechanical yield strength/CVN range*
25	0.20	1.35	0.42	0.016	0.015	0.028	-	0.022	-	-	-	-	0.43	360MPa/70J @ -40°C
20	0.22	1.0-1.6	0.55	0.030 max	0.035	-	-	-	0.3	0.5-0.7	0.2	0.1	0.52	420MPa/60J @ 0°C
20	0.22	1.6	<0.6	0.04 max	0.04	0.003 -0.10	0.003 -0.20	-	-	-	-	-	0.49	450MPa/60J @ 0°C
30	0.13	1.52	0.49	0.005	0.015	0.03	0.10	0.02	0.45	0.72	-	-	0.50	490MPa/>110J @ -20°C

Table 3.5

Typical composition and mechanical properties of thermomechanical controlled processed steel – yield strength range 400 to 500MPa, typical average plate thickness 30mm

Thickness *(mm)*	*Typical composition (by weight %)* *C*	*Mn*	*Si*	*S*	*P*	*Nb*	*V*	*Al*	*Cu*	*Ni*	*Cr*	CE_{IIW}	*Typical mechanical yield strength/CVN range*
30	0.10	1.33	0.28	0.002	0.015	0.027	-	-	-	-	-	0.35	400MPa/190J @ -40°C
32	0.12	1.35	0.30	-	-	-	-	-	0.01	0.02	-	-	398MPa/300J @ -20°C
32	0.07	1.45	0.27	0.001	0.004	-	0.01	0.07	0.19	0.4	-	0.32	400MPa/>300J @ -20°C
30	0.04	1.52	0.22	0.003	0.005	-	-	-	0.60	0.49	0.02	0.37	460MPa/220J @ -40°C

Table 3.6

Typical composition and mechanical properties of quenched and tempered steels – yield strength range from 450 to 1000MPa

Thickness *(mm)*	*Typical composition (by weight %)* *C*	*Mn*	*Si*	*S*	*P*	*Nb*	*V*	*Al*	*Ti*	*Cu*	*Ni*	*Cr*	*Mo*	*B*	CE_{IIW}	*Typical mechanical yield strength/CVN range*
6 – 140	0.18	0.1 - 0.4	0.15-0.35	0.075	0.015	-	<0.02	0.015	<0.02	<0.2	2.25-3.25	1 -1.8	0.2-0.6	-	0.81	550 to 690MPa / 80J @ -84°C
-	0.2	0.1 – 0.4	0.15 – 0.35	0.0254	0.025	0.03	-	-	-	0.25	2.25 – 3.25	1 – 1.8	-	-	0.7	690MPa minimum
30	0.10	1.6	0.50	0.005	0.015	0.03	-	-	-	0.35	0.50	0.15	-	-	0.45	450MPa / >35J @ -40°C
50 – 64	0.12	1.50	0.4	0.005	0.020	-	0.06	-	0.01	0.15	0.30	0.10	-	-	0.43	480MPa / >50J @ -40°C
50	0.11	0.89	0.26	0.003	0.008	0.02	0./01	0.07	0.01	0.15	1.18	0.46	0.38	0.002	0.64	690MPa / >40J @ -40°C
30	0.17	1.2	0.22	-	-	-	-	-	-	-	1.5	0.49	0.5	0.002	0.64	960MPa / >40J @ -40°C

Figure 3.1
Variation in yield strength for 355, 420 and 450 grade steels

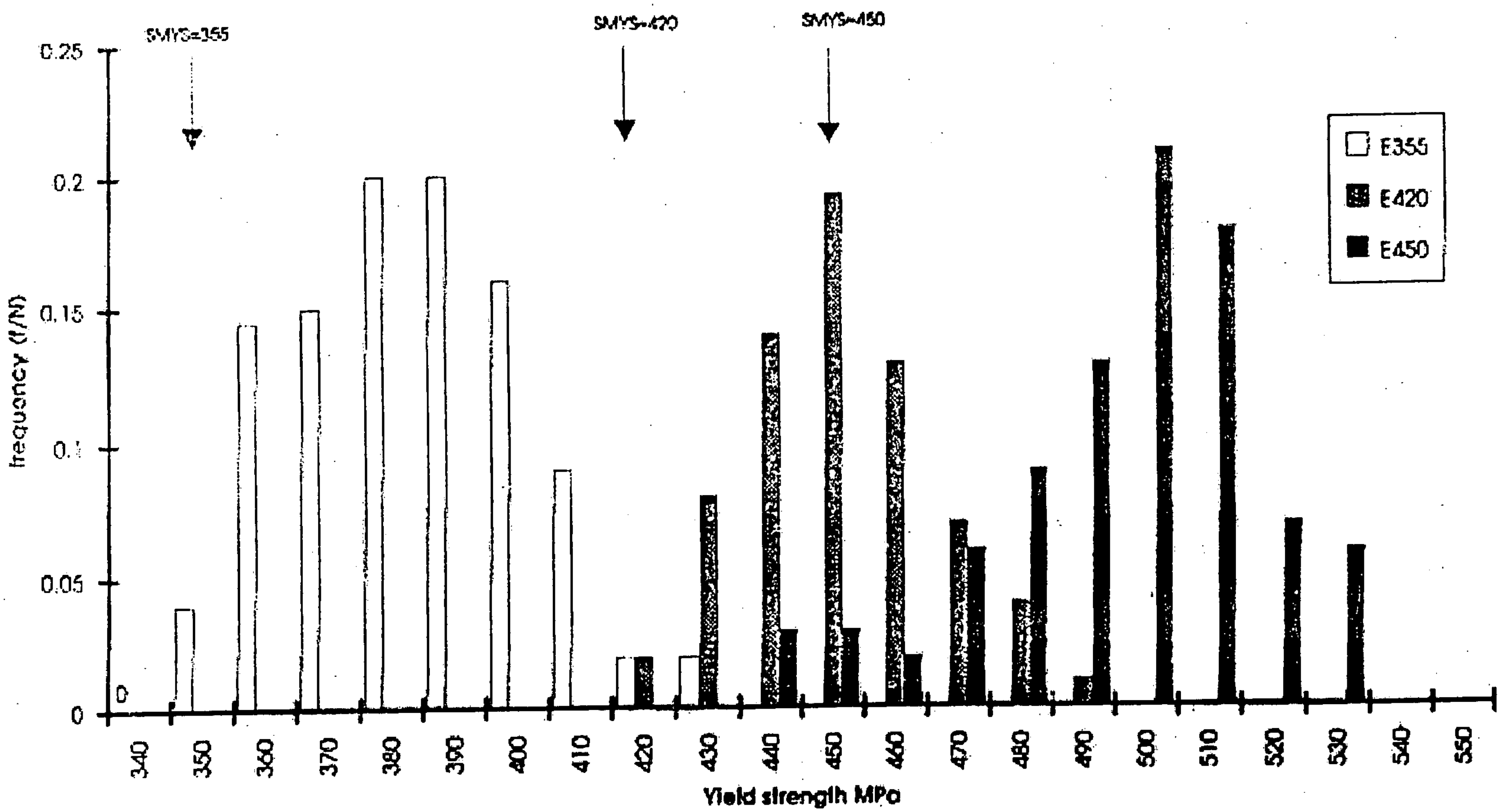


Figure 3.2
Showing typical variation in mechanical properties for a grade 450 steel (35-50mm thick, min σ_y = 430MPa, min UTS = 530MPa, min elongation = 20%, sample size N = 94)

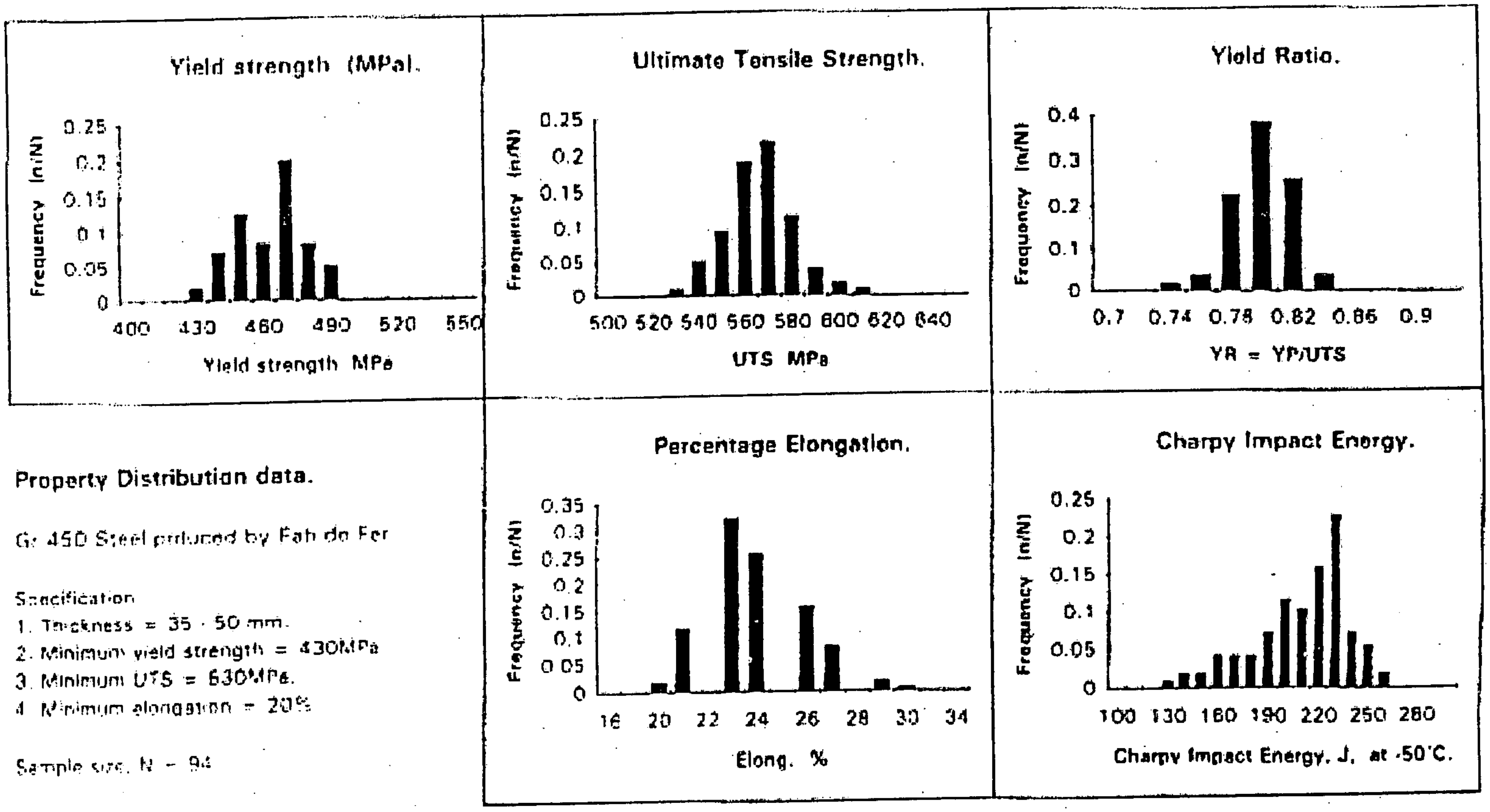


Figure 3.3
Variation in mechanical properties with process route for steel grades 350 and 420 after Denys [3.04]

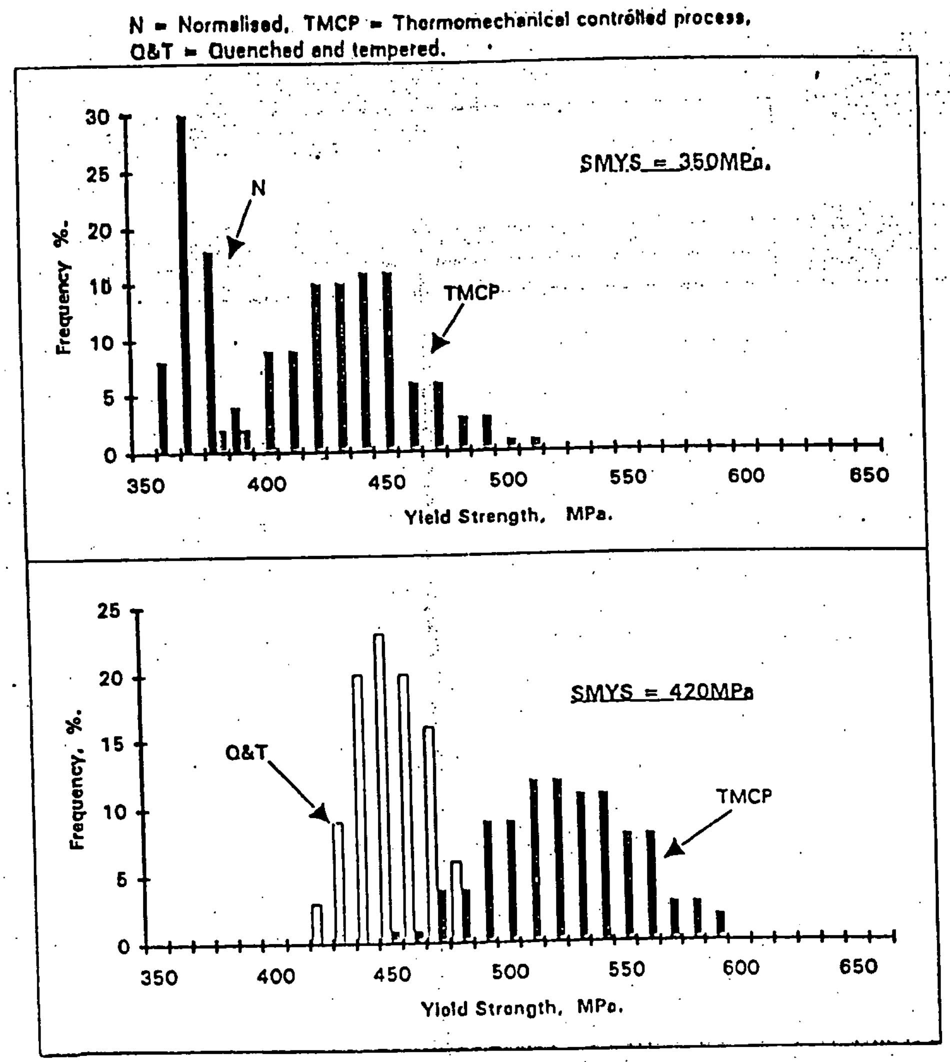
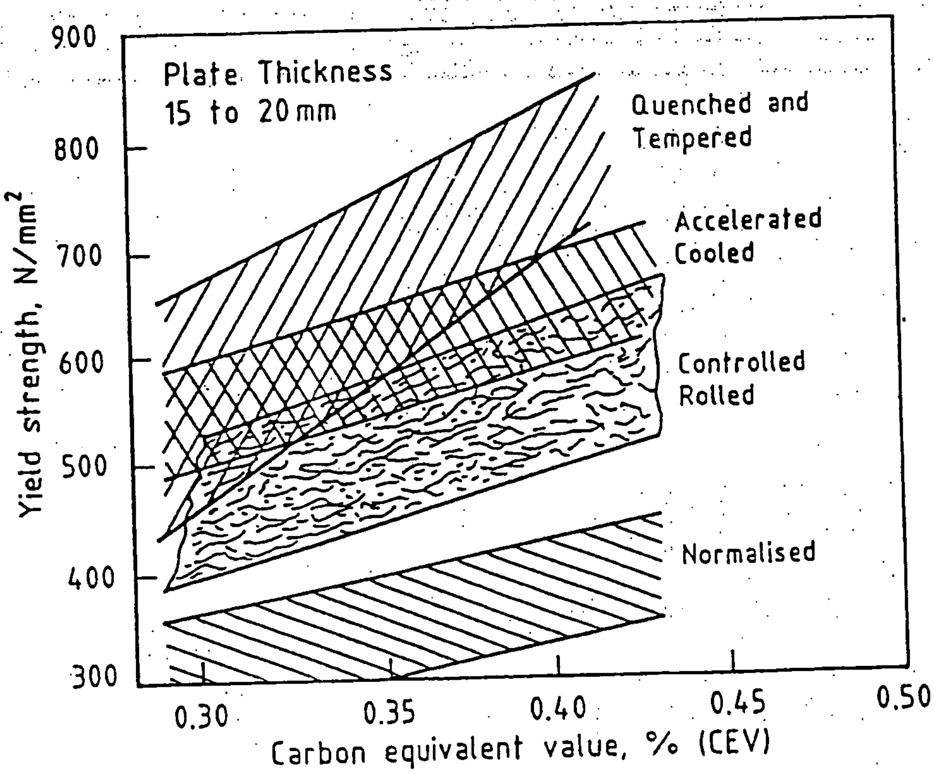
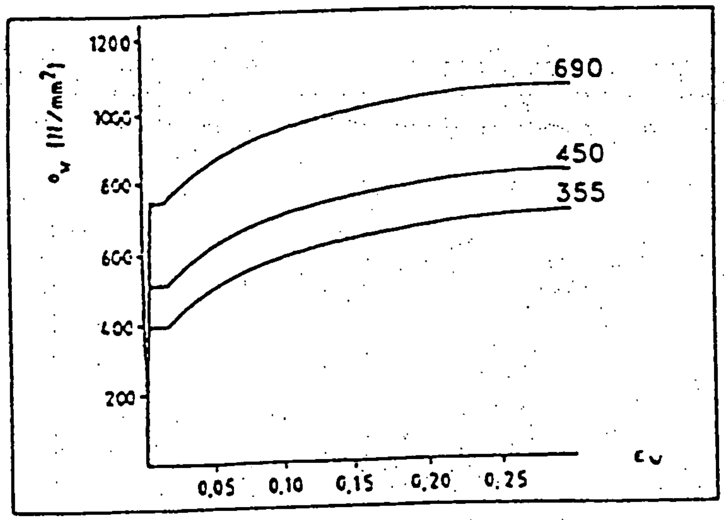


Figure 3.4
Effect of carbon equivalent value and steel processing route on plate strength

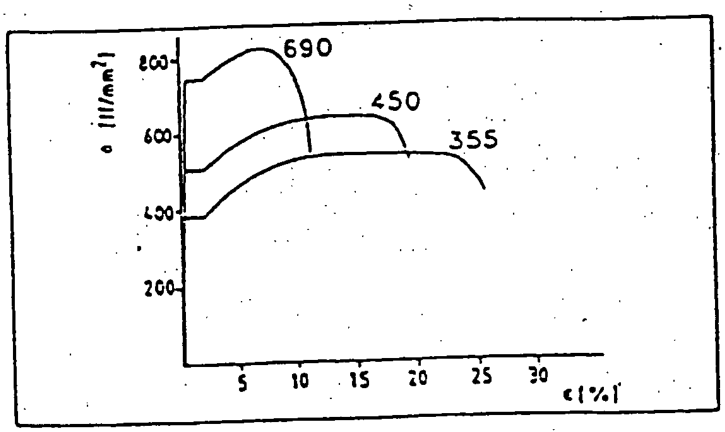


$$CEV = C + \frac{Mn}{6} + \frac{Cr + Mo + V}{5} + \frac{Ni + Cu}{15}$$

Figure 3.5
Typical stress-strain curve for Grades 355, 450 and 690 steel



True stress strain lines for different steel grades



Load-deflection lines for different steel grades

Figure 3.6
Offshore grade steels, nominal thickness 50mm

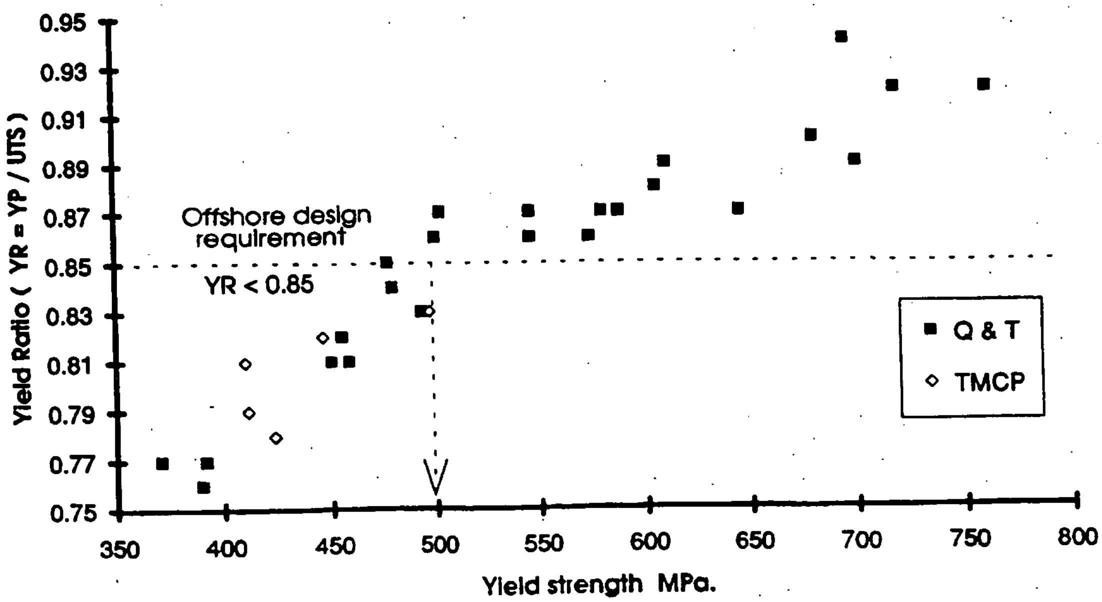


Figure 3.7
Yield ratio of 200 cast and wrought iron high strength steels

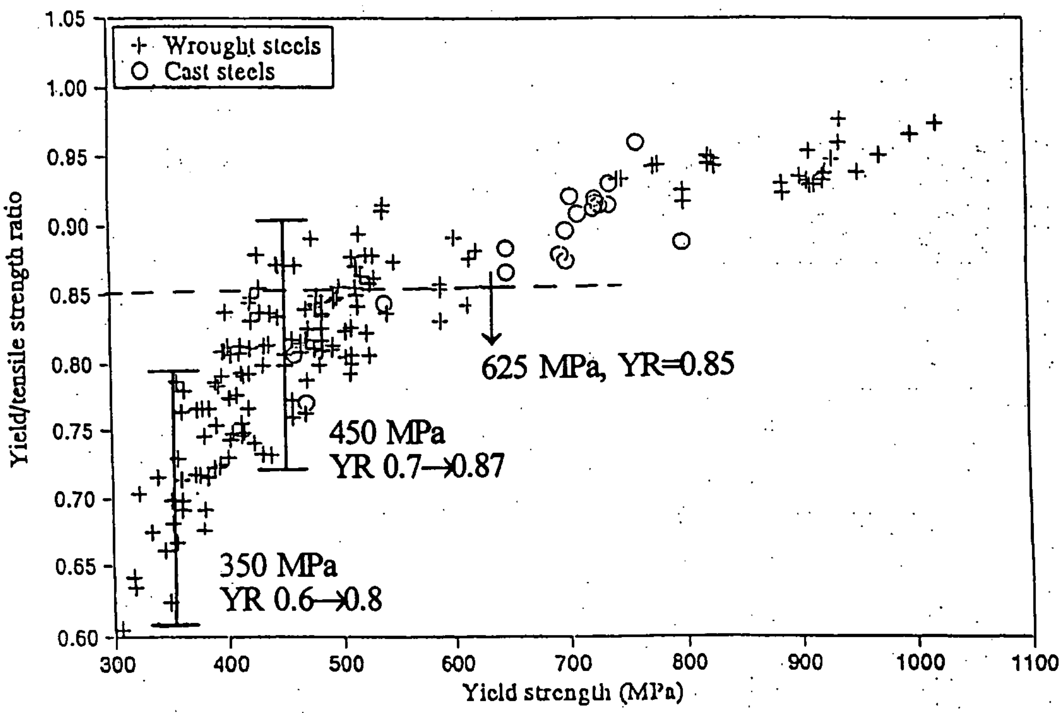
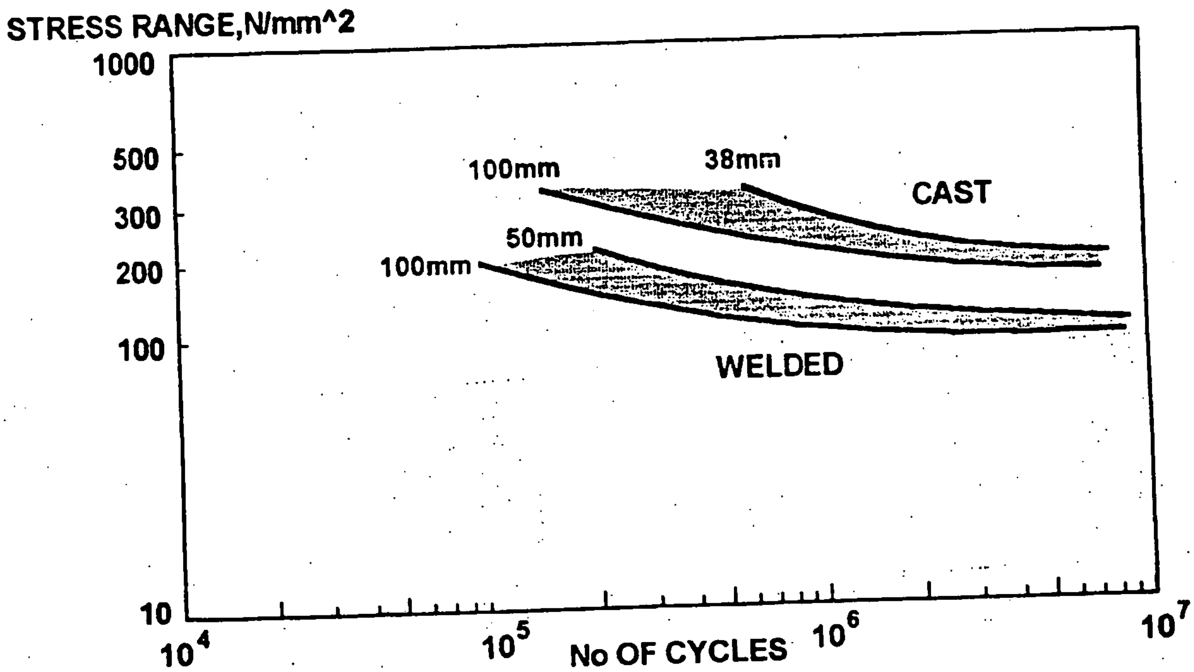


Figure 3.8
Showing comparison of welded and cast steel properties



4. CODES AND STANDARDS

Detailed codes and standards now exist for medium strength structural steels, covering most of the aspects relevant to offshore design. However, for higher strength steels (YS 500-600MPa) the available codes and standards are limited and for even higher strength steels (>600MPa), almost non-existent. This section reviews the available codes and standards, both published and those being developed at present for offshore use. In terms of offshore hazards, table 4.1 shows the current status of codes and standards. The current status of codes and standards will now be reviewed in terms of materials properties.

4.1 ALL PROPERTIES

HSE/D.Energy Offshore Guidance has been developed over many years, with the first edition being published in 1974. The fourth edition was published in 1990, [4.05] with significant amendments being added up to 1995 [4.07]. It is particularly strong in materials properties, performance of structural components etc, and has several sections devoted to high strength steels. In particular the control of hydrogen assisted cracking is addressed in some detail, more than in any other existing code or standard. However as a result of the issue of the DCR Regulations [4.08] in 1996 it has been withdrawn, although it is still available as a document for consultation.

Two ASTM standards [4.01,4.02] cover some aspects of the requirements of high strength steels, although these are limited in their applicability. A808 is concerned with high strength, low ally carbon, manganese steels of structural quality, whilst A514 provides a specification for high yield strength Q&T alloy steels, intended primarily for use in welded bridges and other structures.

The recently published NORSOK standard on the Design of Steel structures [4.09] includes five steel quality levels (DC1 - DC5). However all of these grades are limited to steels with YS equal to or less than 500MPa. For higher strength steels it is stated that the feasibility of such a selection of steel shall be assessed in each case.

The International Association of Classification Societies (IACS) provides a set of requirements for high strength quenched and tempered steels [4.11], for steels with YS in the range 420 – 690MPa, divided into six groups. The requirements include method of manufacture, mechanical properties, and inspection during manufacture.

The DnV offshore standard [4.04] groups steels into three main grades, the highest of which (extra high strength) covers materials with yield strengths from 420 - 690MPa. These grades are linked to impact toughness properties according to weldability requirements, but the improved weldability grade is limited to a maximum YS of 500MPa. A statement is also made that steels with YS> 550MPa shall be subject to special considerations for applications where anaerobic conditions may predominate.

Similarly in the draft ISO Standard for fixed structures (19902) [4.10] steels are classified into five groups, with Grade V covering steels with YS up to 500MPa. It is also stated that further groups may be added when data becomes available. The draft standard includes an important statement on higher strength steels, which is:

'Although steels with yield strengths in excess of 500MPa (73 ksi) are currently available, no agreed standard exists for offshore fixed platform structural use. These are not recognised as offshore fixed platform structural grades and users should take care to ensure that ductility, fracture toughness and weldability will be adequate for the intended application. Attention is drawn to the need to consider fatigue and corrosion conditions, including the tendency for higher strength steels to be more susceptible to hydrogen embrittlement and certain types of stress corrosion. Particular care should be exercised where high strength is developed as a result of alloy additions'.

A separate ISO Technical group is developing a standard for jack-ups, which is expected to include guidance on higher strength steels, appropriate to jack-ups, but has yet to be drafted.

4.2 FATIGUE

New Guidance was published by HSE in 1995 [4.07]. This included a modified set of S-N curves, but these were restricted to steels with yield strength equal to or less than 500MPa, as it was concluded that the test data available were insufficient for higher strength steels. This was particularly true for fatigue in seawater under cathodic protection and free corrosion conditions, where the data available on high strength steel joints were extremely limited. The HSE Guidance recommended that for higher strength steels, data from an approved test programme are used to determine appropriate S-N curves, or fracture mechanics constants. Following incorporation of the DCR Regulations offshore in 1996 the HSE Guidance has been withdrawn.

DnV Rules [4.05] include S-N curves and fracture mechanics constants for steels with YS up to 500MPa.

The NORSOK standard [4.09] provides recommended S-N curves for steels, both in air and seawater. As noted earlier these apply to steels with steel quality levels from I to V, the maximum yield strength being 500MPa. For steels of higher strength it is stated that the feasibility of such a selection shall be assessed in each case.

The draft ISO standard [4.10] states that the limited amount of test data for plate joints with yield strengths up to 540MPa and tubular joints manufactured from high strength steel with yield strengths up to 700MPa suggests that fatigue performance in seawater under CP and under free corrosion is similar to that for medium strength steels, but test data should be used to determine appropriate S-N curves. In addition, the draft standard indicates that for even higher strength steels (700 – 800MPa) the effect of seawater on the fatigue performance of these materials is considered to be more detrimental than for medium strength steels because of their greater susceptibility to cracking from hydrogen embrittlement. In particular, it is noted that several studies have shown that excessively negative CP protection potentials can be a cause of cracking due to the generation of hydrogen which enhances crack growth rates. It is stated in conclusion that it is important that the fatigue performance of high strength steels is understood and that appropriate levels of CP are applied.

4.3 FRACTURE TOUGHNESS

Most codes and standards recommend the need to avoid brittle fracture. Good specifications are published for medium strength steels but generally there is very limited guidance for higher strength steels. Overall avoidance of brittle fracture is based on recommending a minimum value of Charpy energy values according to yield strength. On this basis the International Association of Classification Societies (IACS) [4.11] has recommended for high strength Q&T steels that the average energy from a charpy V notch test should be Re/10 for the longitudinal direction, and 2/3 of this for the transverse direction, i.e. for 690MPa steels (F grade) Charpy energy values of 69J and 46J at a test temperature of -60°C with minimum individual values of 70% of the minimum average, i.e. 48J and 32J respectively [4.11]. Possible limitations on this requirement are considered in section 6.

4.4 HYDROGEN CRACKING

As a result of the detection of cracking in jack-ups in the late 1980s a significant research programme was undertaken on high strength steels which led to new guidance being developed to minimise cracking in practice.

HSE published an amendment to its Guidance [4.07] with a recommendation that the CP level should be limited to a negative voltage no lower than -850mV (Ag/AgCl). To achieve this special measures were recommended, such as voltage limiting diodes to keep potentials within the recommended limits. In addition steels proposed for use offshore in conditions where there is a vulnerability to hydrogen cracking should be assessed using, for example, slow strain rate testing. High strength steels (YS>650MPa) should be examined for the possibility of hydrogen damage in service, both in the

parent material and in the weldments. The HSE Guidance was supported by a published OTH report [4.13] which provided data on the performance of several steels and on the recommended test methods.

The DnV Offshore standard [4.04] also provides guidance on the use of high strength steels in seawater with CP. In this case the recommended range for steels susceptible to hydrogen induced cracking is -770mV to -830mV for steels with YS larger than 550MPa, which is similar to the HSE recommendations.

4.5 DEFECT ACCEPTANCE CRITERIA

BS 7910 published in 1998 [4.06] contains data for calculating crack growth under static and cyclic loading conditions. Recommended values of the constants (A,m) are given for steels with yield strengths up to 600MPa, thus enabling the fatigue crack growth rate acceptance of defects in higher strength steels found during inspection or assumed during design to be quantified. For higher strength steels (>600MPa) it is recommended that test data are required.

4.6 CORROSION PROTECTION

It is stated in the NORSOK standard [4.14] that for high strength steels (YS>700MPa) a special evaluation is required with respect to hydrogen impact. (See EN 10002, metallic materials. Tensile testing. Part 1 method of test).

The DnV code [4.04] provides both general requirements for cathodic protection as well as specific needs for high strength steels. Steels with specified minimum yield strengths >550MPa are subject to special considerations for applications where hydrogen induced stress cracking (HISC) may be anticipated, where qualification testing should be carried out for critical applications such as legs and spud cans. In the absence of such testing to demonstrate that high negative CP levels are not harmful it is stated that the CP level should be limited by the use of special anodes or controlled voltage type (e.g. with diodes) or by other methods. CP potentials levels should also be monitored to ensure compliance with the target range, which is set to be within the limits of -770mV to -830mV (Ag/AgCl). In the case of observed exceedance of this range it is recommended that inspection for HISC should be carried out,

Section 19 of the draft ISO standard [4.10] is concerned with corrosion control, and includes a section on cathodic protection. This states that because of the risks of hydrogen induced stress cracking steels with minimum yield strengths in excess of 720MPa should not be used for critical cathodically protected components without special considerations. In addition, it is stated that any welding or other fabrication affecting ductility or tensile properties, should be carried out according to a qualified procedure, which limits hardness to HV350. It is expected that this will restrict the use of welded structural steels to approximately 550MPa maximum specified minimum yield strength.

For medium strength steels the recommended potential range is -0.8 to -1.1 volts (Ag/AgCl). For some higher strength steels the negative end of this range is expected to be detrimental, in terms of hydrogen cracking etc.

4.7 STATIC STRENGTH OF TUBULAR JOINTS

Current offshore design codes provide equations for determining the static strength of various classes of tubular joints. The strength is generally proportional to yield strength, but data indicate that this proportionality is limited to lower strength steels. As a result the basic equations are limited to steels with YS <500MPa, and there is also a factor to be applied on the ratio of the yield to ultimate strength. This factor varies in different published codes and standards, ranging from 0.67 in API RP2A [4.15] to 0.7 in HSE Guidance [4.05]. The draft ISO standard (19902) [4.10] has increased this ratio to a value of 0.8 for steels with yield strengths up to 500MPa, as a result of new data being available. The background to the application of this factor and its relevance to high strength steels is reviewed in section 3.3.

For higher strength steels the draft ISO standard recommends that the basic ultimate compression capacity equations may be used, together with a ratio of the yield to ultimate strength limited to 0.8, provided adequate ductility can be demonstrated in both the HAZ and parent material (however, the criteria for demonstrating this are not provided). The draft ISO standard highlights that the limit on yield ratio for the tension capacity of joints based on first cracking may need further investigation (see section 3).

The static strength of cracked high strength steel joints is of interest, particularly for the use of flooded member detection. Some recently published data for high strength steels (SE702) [4.16] have been made available from a series of nine static tests performed on large pre-cracked welded tubular joints (six T joints, three Y joints). These were loaded to failure in axial and out-of-plane bending. All specimens had a least one through thickness crack. The results were analysed in terms of both loss of static capacity due to the cracking and by failure assessment diagrams (FAD). The reduction in static strength compared to cracked medium strength steels was about 5% greater, possibly due to differences in crack path (the cracks in the SE702 steel stayed closer to the weld when growing). Using the FAD approach, some discrepancies were found for a T joint with two cracks, giving low values. This was considered possibly due to the inadequacy of the multiple crack correction used. The FAD approach also demonstrated the importance of the fracture toughness value used. For the SE702 steel it was necessary to use the K_Q (nominal) toughness value rather than the maximum toughness value, K_{max}, for which many of the results were unconservative.

4.8 IMPACT PROPERTIES

Low speed impacts can arise from both ship impact and dropped objects. In these cases the ability of the high strength steel to absorb the appropriate energy is one of the main performance requirements. Current codes and standards specify the level of impact energy to be absorbed during ship impact (typically 4MJ) but this is not related to materials properties or yield strength, even for medium grade steels.

High speed impacts can be the result of explosions and the energies of projectiles can vary from several kilojoules to several hundred kilojoules. There is no published guidance on the required material properties and the possible effect of yield strength.

4.9 HIGH TEMPERATURE PROPERTIES

High strength steel components offshore can be subjected to high temperatures as a result of fire, either on the sea or from a jet fire. Unprotected steels can experience temperatures up to 1200°C in a short space of time. Guidance is normally associated with either limiting temperatures (e.g. by passive protection requirements) or by design based on data for lower strength steels at elevated temperatures [4.16]. Some new data have recently been obtained for steels with YS ~450MPa [4.17] (see section 10). For even higher strength steels the data on high temperature performance are extremely limited.

REFERENCES

4.01 ASTM standard A514, 'Standard specification for high yield strength, Q&T alloy steel plate suitable for welding', 2000, ASTM

4.02 ASTM standard A808, 'Standard specification for high strength, low alloy carbon, manganese, columbium, vanadium steel of structural quality with improved notch toughness', 2000, ASTM

4.03 AWS Structural Welding Code, Steel, D1.1-96

4.04 DNV Offshore Standard, Design of Offshore Steel Structures, General (LRFD), 2000

4.05 Dept. of Energy, 'Offshore Installations - Guidance on Design, Construction and certification', HMSO, London, 1990

4.06 British Standard' Guidance on methods for assessing the acceptability of flaws in welded structures, BS 7910, 1991

4.07 HSE, 'Offshore Installations - Guidance on Design, Construction and Certification', HSE Books, 1995, amendment no. 5

4.08 HSE, 'Offshore Installations & Wells (Design & Construction etc) Regulations', HSE Books, 1996

4.09 NORSOK, 'Design of Steel Structures' N-004, 1998

4.10 ISO 'Petroleum & Natural Gas Industries, 'Fixed Steel Offshore Structures, ISO CD 19902, to be published

4.11 International Association of Classification Societies (IACS), 'Unified requirements, Section W16, High Strength Q&T steels for Welded Structures', IACS, London, 1994.

4.12 DnV Offshore standard, Metallic materials, OS-B101, 2000

4.13 *Abernethy, K, Fowler, C M, Jacob, R, Davey V.S.*, 'Hydrogen cracking of legs and spudcans on jack-up drilling rigs - a summary of results of an investigation', HSE Report OTH 91 351, HSE Books

4.14 NORSOK, 'Cathodic Protection', M 503, 1997

4.15 API, Recommended Practice for planning, designing and constructing fixed offshore platforms, APIRP2A 20th edition, API, Washington, USA

4.16 *Talei-Faz B, Dover W D, Brennan F P*, 'Static strength of cracked high strength steel tubular joints', UCL Final report for HSE, 2001

4.17 Steel Construction Institute, 'Experimental data relating to the performance of steel components at high temperatures', Offshore Technology report, OTI 92 602, HSE Books, 1992

4.18 Steel Construction Institute, 'Elevated temperature and high strain rate properties of offshore steels', Offshore Technology report OTO 020 2001, HSE Books.

Table 4.1

Offshore Hazard	Materials Performance Requirement	Codes & Standards for HSS	Comments
Structural Failure	Materials specifications	ASTM standards, A514,A808 [4.01,4.02], DnV Standard [4.04]	DnV code [4.04] covers steel grades up to 690MPa
	Welding specifications	AWS Code [4.03]	Covers steel grades up to 690MPa
	Fatigue of welded joints, members	Limited	Most codes provide a limit of 500MPa on yield strength (YS) for applicability. Specific tests are proposed for high strength steels(HSS) to develop data to support application
	Fracture toughness of steels	IACS recommendations [4.11]	Minimum Charpy values of YS/10
	Hydrogen Embrittlement	HSE Guidance [4.05, 4.07], DnV Rules [4.04]	Control of hydrogen assisted cracking is best described in HSE Guidance Notes [4.04]
	Static strength of tubular joints	Factor to be applied for higher strength steels in several standards, including draft ISO	Modified factor, based on yield ratio (under discussion in draft ISO standard)
	Defect acceptance criteria	BS7910 [4.06]	Data available for HSS (yield strength up to 600MPa)
	Corrosion protection	Limited, HSE Guidance, DnV code [4.04]	Most data provided for medium strength steels
	Inspection & repair	Very limited	-
Boat Impact	Impact performance, large strain capacity	None	Lack of data for HSS
Fire (on the sea, jet fire etc.)	High temperature performance	Very limited	Lack of data for HSS
Blast	High strain rate performance	Very limited	Lack of data for HSS

5. FABRICATION AND WELDING

Most high strength steel applications offshore involve welded fabrication. Forming and welding costs are the most significant item in the cost of a jacket structure, comprising up to 57% of total costs in one reported analysis [5.01]. Improvements in this area are likely to come from welding since cold forming of plates is generally considered to be an efficient and economical process. Welding processes which give greater productivity and/or incorporate a reduction or elimination of pre- and post-welding heating could provide major cost savings, and significant progress has been made. However, as the strength of the steel increases the pre-heating requirement becomes greater as such steels are usually more highly alloyed. If plate thickness is less than 40mm, stress relieving heat treatment is not required for grades with yield strengths up to 450MPa. This can lead to significant time and cost savings in the fabrication procedures. Other areas to consider are the development of welding processes with improved weld deposition rates.

Reports from fabricators [5.02; 5.03; 5.04] indicate that at least up to 450MPa strength levels, welding is no more expensive or difficult for a well organised yard than welding the normal 355 grade. With the highest strength grades, however, more precautions have to be taken.

Weldability of steel is a term that is used to indicate the ease with which sound weldments can be produced using normal welding procedures. The weldment comprises both the weld and the associated heat affected zone. Welding defects such as pores and cracks can be produced as well as undesirable microstructures in the weld and its associated heat affected zone which can lower the resulting mechanical properties of the joint.

Variations in the welding process, such as steel dimensions, weld geometry, heat input and steel composition all influence the resulting microstructure. Nomograms involving thermal severity - joint thickness (mm), heat input of the weld (kJ/mm) and weld preheat required (°C) are often used to indicate the necessary welding procedure to be followed to produce a sound crack-free joint in relation to the particular composition of the steel used which is usually related to carbon equivalent value. In general a steel with lower carbon equivalent value has improved weldability compared to a higher carbon equivalent steel. The two most commonly specified carbon equivalent equations are that recommended by the International Institute of Welding which covers a wide range of steels:

$$CE = CE_{IIW} = C + \frac{Mn}{6} + \frac{Cr + Mo + V}{5} + \frac{Ni + Cu}{15}$$

and the Ito and Bessyo equivalent which is often preferred for modern low carbon steels:

$$CE = P_{CM} = C + \frac{Si}{30} + \frac{Mn + Cu + Cr}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B$$

This latter equation is the one used for high strength steels in the draft DNV Metallic Materials OS-B101 Standard (May 2000). In this guidance document, steels with improved weldability have reduced carbon contents and limitations on the levels of chromium, nickel and molybdenum compared to steels of normal weldability, i.e. they must have reduced CE values.

An alternative approach more commonly used in other parts of the world is the Graville diagram shown in Figure 5.1 which separates the steels into three zones rated by their ease of weldability – zone I easily weldable, zone II weldable with care, and zone III difficult to weld, From this diagram it can be seen that weldability decreases as the carbon equivalent value increases but the diagram also emphasises the extremely important effect of carbon content on weldability. Reducing the carbon content of a steel is the most effective way to improve its weldability.

As the parent strength increases, greater precautions are needed to ensure that welding procedures are satisfactory. The strength increases in the weld are normally produced by alloying since strengthening procedures such as thermomechanical processing cannot be utilised in the weld metal. The welds therefore become more hardenable and precautions are required to prevent weld metal hydrogen cracking. The weldability of modern steels has been greatly improved by their extreme cleanliness, and by their low carbon content and low carbon equivalent values. Low hydrogen consumables are important in reducing the possibility of hydrogen cracking and can also lead to a reduction in the pre-heating requirements. No major problems have been reported in welding steels up to 500MPa yield strength [5.04] in moderate section sizes. At high strength levels, preheating is required and steelmakers are devoting considerable attention to improving the weldability of such steels to try to reduce fabrication costs. For 690 grade steels, for example, preheat temperatures of 125°C are recommended, and electrodes and fluxes with very low hydrogen content must be used in order to prevent hydrogen cracking [5.05].

The formation of hard or brittle phases in the weld HAZ or, indeed, in the weld itself during multi-pass welding, can affect the toughness of the weld and its ability to withstand exposure to hydrogen. Important factors are the grain size in the grain coarsened HAZ near to the fusion line and the microstructural changes that occur in the weld metal during subsequent weld depositions during multi-pass welding. In general, the Charpy toughness of the coarse grained HAZ decreases with increasing heat input and increasing impurity content. Because of this there are often restrictions in the upper levels of heat input that can be used (e.g. 3.5kJ/mm for submerged arc welding) but productivity is not greatly compromised because of the generally thinner sections and smaller volumes of deposited weld metal utilised. Steel that could be welded satisfactorily at higher heat input levels would offer economic advantages [5.06] and recent work [5.07; 5.08] showed that certain steels and weld consumables did satisfy these requirements and could offer further economic advantages.

Published literature indicates that there are weld consumables which can produce the necessary material properties required in service, even for the 690 grade steels [5.09]. However, at the highest strength levels envisaged there is much less experience and availability of weld consumables with suitable properties, particularly in respect of toughness. In addition, significant pre-heating and interpass control are necessary in order to avoid hydrogen cracking problems. Weld metal microstructures are determined primarily by the chemical composition, the amount of non-metallic inclusions present in the microstructure which affect phase nucleation, and by the cooling rate. Alloy design aims to maximise the amount of acicular ferrite present and to minimise the effects of undesirable microstructures such as coarse grain size, grain boundary ferrite and coarse martensite/austenite/carbide constituents (MAC).

The welding consumables employ sophisticated alloying techniques, incorporating the optimum balance of deoxidising elements (aluminium, silicon and manganese) to produce a high density of small non-metallic inclusions which are known to act as intragranular nucleation sites for acicular ferrite. The carbon content is generally kept low to aid weldability, so the increased strength is achieved through additions of molybdenum in SAW wires and titanium-boron in FCAW wires, and the impact toughness is improved with nickel additions. In the Cranfield study [5.08] consumables up to 550MPa yield strength showed adequate toughness throughout the weld, with upper shelf values >150J and the 50J impact transition temperatures below -60°C. All welds had low hardness values and showed no indication of hydrogen cracking. Acicular ferrite was the major microstructural feature of the welds and microstructures generally coarsened as heat input increased.

For both SAW and FCAW, 690MPa consumables showed mixed microstructures containing acicular ferrite, martensite and polygonal ferrite. Impact properties were inferior to the 550MPa welds with upper shelf values ranging from 80-100J and 50J impact transition temperatures between –50 and -80°C. It was concluded in this study that more development work is needed before these consumables can be specified generally for offshore application.

In most welded structures it is considered desirable to overmatch the yield strength of the parent plate and the related HAZ. This is because if the welds undermatch then any enforced deformation will be concentrated in the relatively small weld metal volumes leading to high strain in these zones. If such zones have reduced toughness values, which is generally the case at the highest strength levels, then there is an increased likelihood of failure. Weld metal specifications often call for 20 to 30% overmatch which is easily achievable in 450 and 550 grades with satisfactory weld property performance. The problem arises in producing the necessary combination of properties in the weld metal required at the highest strength levels, i.e. 690 grade. In other applications such as storage tanks, undermatching welds have been used in high strength steel structures which have generally performed satisfactorily because the weld metals have good toughness. The normal statistical variations in yield strength that occur in steel plate (discussed in Section 3), also occur in weld metals. This poses an additional problem because there is a distinct possibility that, unless significant levels of overmatch are specified, the lower strength weld metals will undermatch the highest strength parent material in particular project fabrication programmes, leading to certain joints being undermatched. The importance of this effect is provoking considerable interest at the moment, particularly with regard to the higher grade steels.

It has been reported that the residual stresses in restrained high strength steel joints are less than those encountered in lower strength steels which can have an important influence on subsequent fatigue and fracture behaviour. Bennett et al [5.05] claimed that the residual stresses in a 40mm thick restrained joint were only approximately one half of those obtained with mild steel and were largely independent of heat input. The explanation of this effect was thought to be partial counterbalancing of the thermal contraction stresses by the martensitic/bainitic transformation which occurred during cooling. The potential benefit of this different behaviour seems not to have been utilised significantly to date. Further work is needed to provide confidence in this approach.

A recent report details the welding practices and procedures used for welding the Elgin jacket [5.09]. High strength steels, Superelso 500 and Superelso 600, were used on the project in leg chords. These materials have specified minimum yield strengths of 450MPa and 550MPs and yield ratios of 0.78 and 0.80. Significant use was made of MMA welding during the fabrication because of its fully positional welding on site capability. Oerlikon electrodes, Tenacito 70, were used which gave yield strengths between 490 and 550MPa and excellent Charpy V-notch toughness values of 130J at -40°C.

The welding of the chord to the prefabricated 160mm thick rack sections was carried out with a minimum preheating temperature of 150°C followed by a dehydrogenation treatment of 2 hours at 200°C. Very low repair rates (<1%) were reported for the project.

REFERENCES

5.01 *Webster S,* 'Structural materials for offshore structures – past, present and future', IBC Conference – The Safe Design and Fabrication of Offshore Structures, Sept 1993, publ IBC

5.02 *Lessels J, Rohde W, Pontermoli M, and Devillers L,* 'The present state of technical knowledge of offshore structural steels and future material requirements', May 1992, CEC 7210.ZZ466

5.03 Private communication, F. Forster, AMEC

5.04 *Rogers K T, Thornton C E, and Naylor K D,* 'Progress in the welding of higher strength steels for offshore applications', OMAE Conf. Vol.III, 1989, 309

5.05 *Bennett W T, Cadiau L, and Caudreuse L,* 'Steels for jack-up legs' in Recent Developments in Jack-up Platforms, ed Boswell L F, and D'Mello C, 1992

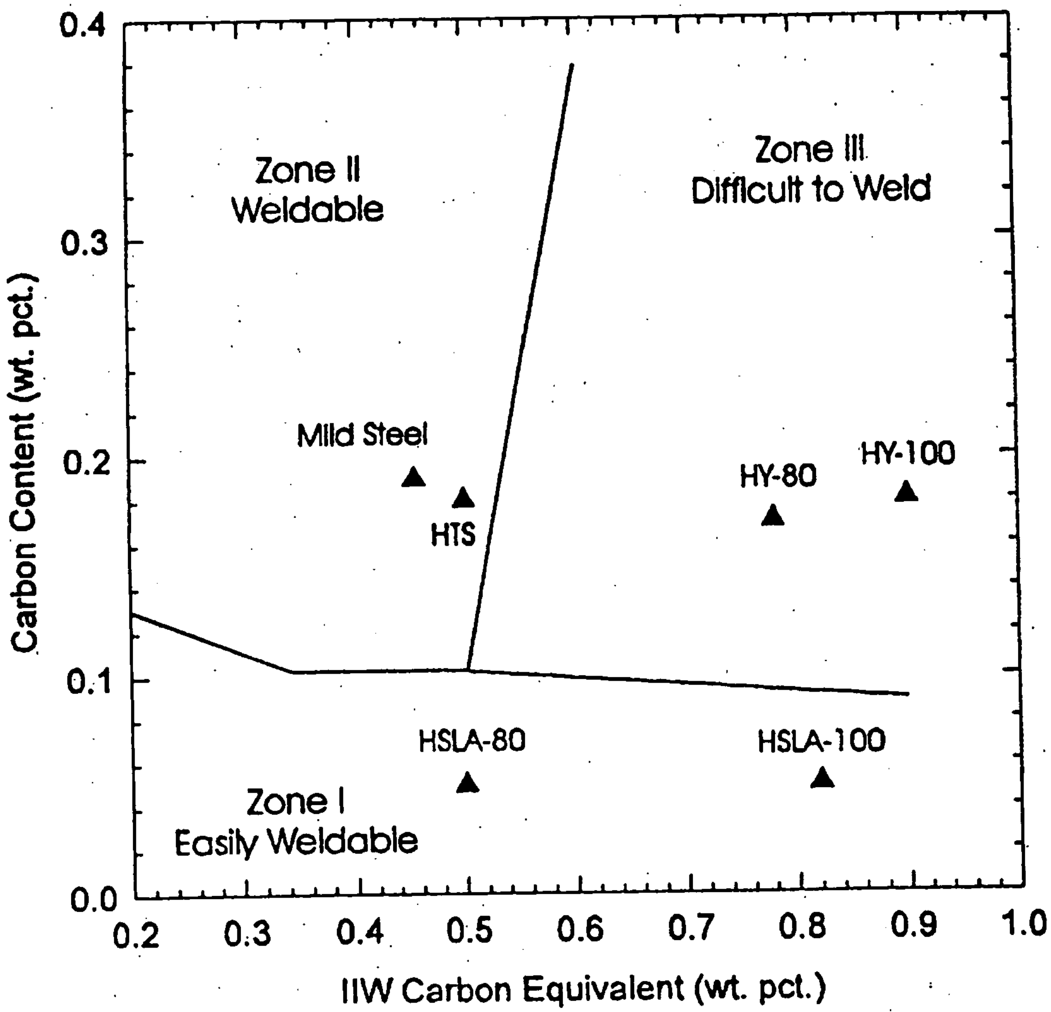
5.06 *Krancioch J J,* 'Fabrication of offshore structures and potential productivity increases', The Safe Design and Fabrication of Offshore Structures, Sept 1993, Publ. IBC

5.07 High Strength Steels in Offshore Engineering, MTD Publication 95/100, ISBN 1-870553-21-7, 1995

5.08 *Billingham J, Blackman S, and Norrish J,* 'Further assessment of high strength weld metals for use in offshore engineering applications', Cranfield Final Report, January 1998

5.09 *Bews R,* 'MMA Welding Gas from Strength to Strength' Welding and Materials Fabrication, July/August 2000.

Figure 5.1
Criteria of Steel Weldability – Cracking Susceptibility



6. TOUGHNESS

Toughness may be loosely described as a measure of the resistance to failure in the presence of a crack, notch or similar stress concentrator. High toughness therefore is generally recognised as a desirable property for offshore steels.

A high toughness material is one where a considerable amount of plastic deformation is required at the crack tip before the crack can be made to advance. Conversely, if the application of stress causes the elastic failure of atomic bonds at the crack tip, relatively little energy of deformation is involved, and the result is a brittle fracture.

The word 'toughness' is used for two quite separate quantities. They are more correctly described as 'Impact Toughness' and 'Fracture Toughness'.

Impact toughness is an energy measurement (Joules, or ft-lbs) and commonly relates to the Charpy V-notch test. Fracture toughness is a calculated value for the critical stress intensity factor ($N.m^{-3/2}$ or $MPa\sqrt{m}$ or $psi\sqrt{i}$) assessed for ductile materials from crack-tip-opening-displacement (CTOD) tests or J-integral tests.

Relationships between these quantities are empirical. The relationships have been well validated over many years for structural steels in moderate section thickness. This has permitted the more readily available Charpy impact data to be used as an indicator to the adequacy of the fracture toughness.

Where the same relationships between impact testing and fracture toughness are extended to thick-sectioned material and to high strength steel, specifications should be viewed with caution until it has been demonstrated that adequate factors of reserve are incorporated for the new conditions. Direct testing for fracture toughness may be preferable, particularly since it can reduce some of the uncertainties related to the effect of material thickness.

In ferritic steels, the fracture toughness is affected by temperature, by strain rate and by geometry. The latter influence is also known as 'the stress state', 'the degree of triaxiality' or 'the thickness effect'. The apparent changes in toughness that result from the geometry are not quantified at all by the Charpy test, which always uses a standard small (10mm thick) specimen. Despite this, Charpy results are widely used in materials selection and in current codes and standards.

6.1 DUCTILE TO BRITTLE TRANSITION

Both the impact toughness and the fracture toughness of ferritic steel are characterised by a ductile-to-brittle transition as the temperature is reduced. This corresponds with a change in the mechanism of crack movement, from plastic blunting and plastic tearing (ductile control) at the higher temperature to cleavage (brittle fracture) at the lower temperature. The transition occurs over a relatively narrow range of temperature, typically 30°C, but often involves considerable experimental scatter. As a result, there are several different definitions of the transition temperature from the same data. Low transition temperatures and high 'upper-shelf' values of toughness are seen as beneficial.

The transition temperature (TT) is not an invariant property of a given material, even for a fixed composition, grain size etc. The TT varies with the state of stress (which means that it depends on the size and geometry that has been used) and rate of loading. Increasing the thickness of the specimen, or increasing the rate of loading, produce an increase in the TT.

Without these complications, it would be relatively simple to avoid brittle fracture - the basic requirement would be to select material for which the TT was below the service temperature range. It would still be necessary to take into account that fabrication processes such as welding affect the

material and its TT, both as a result of changes in the material from the thermal cycling and from the introduction of flaws (that have the effect of increasing the TT). It would also be necessary to ensure that the results related to the correct environmental exposure.

Unfortunately, the commencement and the extent of plastic deformation at the tip of a crack are significantly affected by the geometry. At a particular temperature, a thin-section low-speed fracture mechanics test may exhibit ductile behaviour whereas a thick-section test may give a brittle fracture, even where samples have been cut from exactly the same block of material. In structural applications, a known thickness of material may be selected, but complex joint geometry or complex stress fields may again influence the balance between ductile deformation and brittle fracture at a crack tip. Operating at a temperature above the ductile-to-brittle transition temperature therefore does not automatically guarantee the avoidance of brittle fracture in a structural component.

6.2 CHARPY V-NOTCH VALUES AND HIGH STRENGTH STEEL

The Charpy V-notch impact test is probably the best-known of several small-scale tests designed to study the resistance of the material to impact loading in the presence of a standardised stress concentration, by recording the energy absorbed from the pendulum by the specimen. Charpy results however cannot be considered to be directly relevant to structural behaviour.

Much of the early experience with Charpy specifications relates to steel that was produced by the normalisation route, which gave a fine-grained ferrite and pearlite microstructure, but there was some risk of producing the more brittle martensite microstructure in the heat affected zones of welds. By demonstrating that the product avoided brittleness in a low temperature Charpy test (by absorbing at least 27J of energy for example), the implication was that those microstructures that were most at risk of undergoing fracture at the operating temperature were absent. The speed of the impact test was regarded as 'severe' and likely to promote brittleness. The influence of thickness (geometry, stress state) on the transition temperature was not overlooked, but it was addressed by means of applying a temperature shift. Instead of seeking a certain minimum Charpy energy at the design temperature, these values were required at a lower temperature that was adjusted according to the thickness of the material in the structure.

The offshore industry applied this experience to normalised steel with a SMYS of 355MPa and sought improved toughness levels. Steels made by various thermo-mechanically controlled processing (TMCP) techniques were introduced because of their improved weldability. Products with SMYS around 450MPa were produced specifically for offshore structural use and generally exhibited excellent impact toughness and low transition temperatures (frequently below -80°C). Higher strength steel is usually manufactured using the quench-and-temper (Q & T) route. The quench produces a finely-structured martensite or bainite product that is usually stronger than required and (except in very low carbon steels) is usually too brittle for direct use. The tempering reduces the strength, relieves some of the residual stress and improves the impact toughness. More recent developments in the processing try to avoid the intermediate production of very brittle phases.

For simplicity, the design assumption that higher strength steels carry a proportionally higher load led to the 'R_e/10' requirement so that the required longitudinal impact toughness in Joules is equal to one tenth of the SMYS in MPa. The required transverse value is usually taken as two thirds of this to allow for anisotropic effects in the microstructure.

It is not clear, however, whether the same temperature offset between the design temperature and the impact test temperature is equally applicable to the higher strength steels. These comprise different microstructures, finer grain sizes, differing degrees of scatter and much higher upper shelf energy levels than the materials that were originally involved in the validation tests. Figure 6.1 shows Charpy data for four modern high strength steels from one manufacturer. It illustrates the differences between these steels in terms of the upper shelf Charpy value and the steepness of the ductile-to-

brittle transition temperature. The transition temperature also varies considerably, and the graph shows the difficulty in defining this value for some of the steels.

A 1966 appraisal of the use of high strength steels in offshore installations addressed the properties of offshore steels with a yield strength of 450MPa and above and jack-up steel data in some detail [6.01]. The paper includes a histogram of fusion line Charpy data at -40°C for Grade 450EMZ steel for weld heat inputs of 0.8kJ/mm and 3.0kJ/mm to illustrate that good impact toughness levels can be achieved in high strength steels. About 80% of the results absorbed more than 150J of energy. Comparable fracture toughness data give about 85% of CTOD results above 0.5mm, which should imply very acceptable defect tolerance levels. In addition, Charpy scatter-band data for modern, low carbon, low alloy, roller quenched and tempered (RQT) steels are contrasted against those for older versions of the same material to show the improvements from modern chemistries and production. Between +20°C and -60°C, absorbed energy levels had been increased by approximately 2½ times as a result of modern practices. A selection of impact data for heat affected zone and steel plate with yield strengths to 765MPa is included in the paper. Most of the steels are in the yield strength range 450 – 580 MPa although there are limited results from the higher strength steels.

6.3 FRACTURE MECHANICS VALUES

Fracture mechanics values of toughness are linked to the size of flaw in the structure that is critical under the applied stress. In one of its simpler forms, the relationship is expressed as

$$K_C = \alpha . \sigma_C \sqrt{\pi . a_C}$$

where K_C is the fracture toughness, α is a geometry correction, σ is the structural stress, a is the parameter that relates to the crack size and the suffix C indicates critical conditions for the initiation of crack movement.

Reserve factors are usually applied to set limits for repair of detected flaws before this critical size is attained. In addition, flaw growth rates by sub-critical mechanisms (such as fatigue, stress corrosion, etc) are used in conjunction with flaw detection methodology to determine a commensurate flaw inspection schedule. Fracture toughness therefore is an important aspect of material selection, and one that is extensively incorporated into codes and standards for medium strength steels.

Fracture testing can be performed on the full thickness of the structural material, but uncertainties will still arise where the geometry is complex and where the residual stresses are difficult to define. Environmental influences also must be taken into account.

The user needs to be clear on the terminology that is used in fracture toughness. Lower-bound values for toughness are identified as K_{IC}. If there is an environmental influence, for example producing stress corrosion cracking, the toughness may be further reduced to K_{ISCC}. The use of the Roman I in these formats signifies two important facts about the value – that plastic deformation is effectively at a minimum and that it is 'opening mode' loading (modes II and III exist, but mode I values are usually the lowest).

K_{IC} is called the 'plane strain fracture toughness' because the crack tip behaviour is dominated by the elastic loading condition of plane strain and this has the effect of restricting plastic deformation. A standard fracture mechanics specimen usually has a geometry in which the stress state is determined by the thickness (B). Plane strain dominates when B is large. Theoretically, the minimum toughness relates to infinitely thick specimens, but the engineering approximation K_{IC} is given where the specimen thickness satisfies the inequality

$$B \geq 2\tfrac{1}{2}\left(\frac{K_{IC}}{\sigma_{YP}}\right)^2$$

where σ_{YP} is the yield or proof stress.

K_{IC} will change when the temperature or strain rate is altered. The disadvantage of using K_{IC} for calculations when thinner sections are used is simply that the values may be excessively conservative, requiring applied stresses to be limited, or very small cracks to be repaired.
If a toughness value has been obtained from a sample that is not thick enough to qualify as plane-strain-dominated, the toughness is written as K_C. It is then called the 'fracture toughness', but it is important to realise that this number now also depends on the thickness that was used in the test. [Figure 6.2] K_C will also change when the temperature or strain rate is altered. It is potentially dangerous to measure the fracture toughness in a 'thin' specimen and use this toughness value in the design of a structure that is made from thicker material.

Standardised fracture mechanics specimens are designed to produce a high degree of constraint in order to promote lower-bound values. Fracture toughness tests relate to initiation of crack movement and such tests may detect locally arrested cracking as 'pop-in' events when the load-displacement trace shows a discontinuity. It then may be debatable whether the initiation event is significant, i.e. whether a similar crack initiation would have continued to propagate in the structure. Guidance on interpreting pop-in events from tests may be found in BS 7448:Part 1:1991 [6.02], whereas Part 2 deals specifically with fracture toughness evaluation for welds [6.03].

6.4 FRACTURE TOUGHNESS OF HIGH STRENGTH STEELS

In general, recent modern steel-making developments for structural steels that produce ultra-fine grained low alloy products have much-improved upper-shelf toughness and significantly lower transition temperatures for the same thickness compared with the older products of comparable strength.

Refining the grain size is the best option for increasing the strength, because toughness is improved at the same time, but grain growth controllers are required in weldable steels. Increasing the alloy content of steel to increase its strength tends to reduce toughness, hence steel-makers offset this effect by grain size control during the manufacturing.

Weld metals for use with high strength steel are required to show comparable levels of strength and therefore have to develop strong tough structures on solidification. Ultra-high strength steels are also likely to rely more heavily on increased levels of alloying to achieve the required strength, as an economic limit for grain refinement is reached. Very high toughness, therefore, may be less readily achieved in these two cases. Selection should therefore be based on adequate levels of toughness for the purpose, rather than absolute values.

Local precipitation changes in the HAZ regions of welds, particularly in more highly alloyed steels, may produce scatter in fracture toughness values. Regions causing this scatter have been called 'local brittle zones' or LBZs but this term also applies to effects caused by local grain coarsening.

Recent weldability tests on Thyssen steels up to 690MPa utilised Charpy and CTOD tests [6.04]. Welded high strength steels exhibited satisfactory performance to 690MPa, with TMCP and accelerated cooled steels being especially good. For Q & T steels above 690MPa, it was suggested that high weld heat input should be restricted and fracture toughness should be checked.

Studies were performed in the early 1990s for jack-up steels [6.05] by Creusot-Loire Industrie, after Friede and Goldman Ltd pointed out that steel specifications and production techniques had evolved over time for that particular application. The pioneering jack-up designs used 275-350MPa material

for chords and 205-250MPa for bracing members, whereas these had evolved to 520-690MPa and 345-550MPa in the 1980s and were predicted to rise to 750-900MPa and 550-690MPa respectively. The Creusot-Loire steel (A517Q mod) had a typical actual yield stress of 770MPa and YR of 0.91, with Charpy levels of 70J at -60°C. Welding preparation was regarded as very important. Correlation between Charpy and CTOD (δ) were investigated. CTOD tests gave values of 0.15mm for δ_u at -20°C for 180mm thick plate and HAZ, and the steel met or exceeded the criteria of the jack-up industry.

Although thick sections are capable of developing plane strain conditions making it difficult for plasticity to occur at the crack tip, it does not mean that plane strain conditions always apply in thick sections. A surface crack that is long but not very deep may not experience constraint of yielding, hence its apparent toughness may be significantly above K_{IC}. In addition, when thick sections are assessed by standard fracture toughness tests, the crack front in the specimen may be appreciably larger that that of any tolerable flaw in the structure. It therefore may be argued that standard tests where cleavage occurs produce pessimistic values. Code requirements for applications such as racks in jack-up structures, where yield strength is typically 690MPa or above, may need to take such factors into consideration to avoid being unduly conservative [6.06].

The 1996 paper by Stacey, Sharp and King [6.01] includes some CTOD data for a Creusot Loire A517F steel (736MPa) and a Nippon Welten 780 (765MPa) steel. These were tested at -15°C both as 30mm fracture toughness samples and in their full thickness, which ranged from 170mm to 250mm. For the 18 large-scale tests, the CTOD values reached at least 1.00mm (which was the nominal capacity of the test facility) in all cases, whereas the 30mm samples gave CTOD result ranges of 0.22 – 0.81mm and 0.15 – 0.27mm respectively. This is apparently contradictory to the effect of increased thickness. However, the lower results from the smaller specimens were attributed to the influence of specimen size on the inhibition of the plastic deformation in ductile process of slow stable tearing. In the DnV tests that were being reported [6.07], only maximum load values were returned from the full-sized specimens (from the few tests that attained this before exceeding the machine capacity), whereas a detected instability in crack tip tearing in the smaller specimens requires CTOD calculations based on that event. An apparent complication arises from comparing δ_u and δ_m results and the work may also reflect the difficulty of detecting small failure events in large-scale specimens.

In work [6.08] at DERA, comparisons were made among the ductile-to-brittle transition temperatures from three different tests:

(i) conventional Charpy V-notch samples (a/W = 0.2),
(ii) Charpy samples where constraint had been increased by extending the notch with a narrow spark-eroded slot (a/W = 0.45), and
(iii) full-thickness J-based tests (a/W = 0.3) generally on 50-60mm thick samples.

The main series of tests comprised 3%Ni Q & T steels or boron-treated Q & T steels with yield stress in the range 550-700MPa, but the work included a 600MPa weld metal and a 300MPa C-Mn steel. The comparisons of Charpy-sized tests showed that the C-Mn steel and weld metal have relatively little notch acuity shift, whereas this could be up to 50°C for high strength steels, and up to 80°C between Charpy and full thickness dynamic tests. Attention was drawn to the practice of requiring Charpy tests at only -40°C for selection of high strength steel at design temperatures around 0°C [6.09]. It was concluded that reliance on the conventional Charpy test for fracture avoidance could be dangerous for high strength steels and weld metals, especially if the criteria were based on those for lower strength C-Mn steels without recognising the possible differences in behaviour.

For ultra-high strength steels, alloying additions are likely to raise the transition temperature. Hence, it may be unwise just to rely on upper-shelf fracture toughness values from a particular test temperature (related to the design temperature) even if these toughness values seem very high. It is more prudent to investigate where the transition temperature actually lies in relation to the test temperature. The same doubts on the temperature offset apply to the nominal correlation between fracture toughness and a required energy level in Charpy tests (27J, 'R_e/10' etc). In assessing such

correlations, it should be borne in mind that mechanical properties such as the actual yield stress frequently vary considerably from batch to batch, according to the production process and the manufacturer, and this will affect the amount of crack tip plastic deformation. When such influences need to be assessed, a clearer picture is likely to emerge if Charpy and fracture toughness test results from the same batch are used in the correlation. Repeat tests should be done as required on other batches to determine the extent of scatter from variations in materials properties.

6.5 FLAW ASSESSMENT CONSIDERATIONS FOR HIGH STRENGTH STEELS

The main purposes for performing fracture mechanics tests are to determine at the design stage how big a flaw would cause a problem, and after construction whether a detected flaw needs to be repaired. These are based on the calculation of the critical effective crack length parameter, $(\bar{a}_{eff})_{CRIT}$ from fracture mechanics. Reserve factors give the tolerable flaw sizes.
The growth of the flaw from mechanisms including fatigue, corrosion fatigue etc must be considered. Knowledge of the predicted flaw growth rates allows sensible inspection intervals to be fixed. Calculations must check that the tolerable size is not exceeded before any necessary remedial action can be carried out. At the design stage, this is based on the largest size of flaw that could escape detection. In service, it is based on the current size of a detected flaw.

The procedures for measuring the fracture toughness and for calculating the tolerable flaw size parameter are given in BS 7448:Part1:1991 'Fracture mechanics toughness tests : Part 1 : Method for determination of K_{IC}, critical CTOD and critical J values of metallic materials' and BS 7910:1999 'Guide on methods for assessing the acceptability of flaws in fusion welded structures'[6.02; 6.10]. A separate standard now deals with the determination of plane strain fracture toughness values [6.11].

Taking **critical conditions in plane stress** but dropping the suffixes for clarity here, it may be shown that the basic relationships give

$$\frac{K^2}{E} \approx J \approx (\sigma_{YP}.\delta) \quad \text{and} \quad \frac{K^2}{\pi.(\sigma_{app}{}^2)} = \bar{a}_{eff}$$

If using CTOD, it is important to notice that it is the critical value of $(\sigma_{YP}.\delta)$ that relates to fracture toughness K_C and not just the displacement δ_{CRIT}.

When a designer contemplates moving to a higher strength of steel, it is not immediately obvious whether the toughness requirement also need to be increased, should stay the same, or be relaxed. It is affected by several related factors. The stronger steel is usually chosen to allow weight reduction, so that less steel supports the same load, but the design stress is often increased proportionally (to be a set fraction of the yield stress). This gives maximum weight-saving, but there are two disadvantages; the toughness must show a corresponding increase to prevent the critical flaw size going down, and also the new steel must be more resistant to crack growth mechanisms such as fatigue because the working stress will have increased. Although modern high strength steels do have improved fatigue resistance compared with conventional offshore steels now in service, it is usually not commensurate with the increase in strength.

An alternative approach, currently under-used, is to design on the basis of the improved resistance to crack growth. This should permit higher structural design stresses than currently employed, with some reduction in weight from reduced section thickness. The improvement implies working with the higher strength steels at a design stress that is a lower fraction of the yield stress. Further guidance on this is given in Appendix 6.

Despite the above complications in obtaining the fracture toughness value that characterises the particular geometry and loading conditions, fracture mechanics is an extremely useful tool in practice. It is relatively simple to obtain fracture toughness values that are predicted to be conservative in use.

The degree of conservatism may be open to debate, but the information can be used

(i) at the design stage (to evaluate tolerable flaw sizes related to the difficulty of inspection and reliability of the inspection techniques), [6.12],
(ii) during service (to assess the significance of any crack growth and to set inspection periods) and
(iii) as part of the contingency planning (regarding the stability of cracks following accidents).

It is therefore advisable to derive the toughness requirements from the flaw acceptability limits for each structure using recognised procedures and to perform a sensitivity analysis on the outcome.

A common European structural integrity assessment procedure (SINTAP) is being developed. because of the diversity of fracture mechanics data [6.13] and the need of industrial users to find a reliable correlation with Charpy impact energy data [6.14]. The methodology uses a fracture toughness parameter K_{mat} to characterise the particular material, and a probability distribution $P\{K_{mat}\}$ that enables the confidence level of the assessment to be quantified. Where Charpy data are being used, lower-bound correlations are used for lower shelf and upper shelf behaviour, and a 'master curve' correlation based on statistics is used in the transition region. The procedure uses the 27/28J Charpy transition temperature, the $100MPa\sqrt{m}$ fracture toughness temperature and K_{mat} for 25mm thick specimens, with formulae to correct for the design temperature and the appropriate structural dimensions. For confidence in data input, the verification calculations show that the results from six tests should give a 75% probability of having a conservative mean fracture toughness value [6.15]. However, the procedure is not being written specifically for high strength steels with SMYS above 450MPa. The procedure appears promising [6.16], but more validation may be needed for the stronger steels. There has been a contemporary study of the fracture properties of ship steel plates [6.17], and renewed interest in statistical assessment of Charpy data [6.18].

6.6 SUMMARY OF TOUGHNESS CONSIDERATIONS

The word 'toughness' is used for two quite separate quantities, the 'Impact Toughness' (Charpy absorbed energy value) and the 'Fracture Toughness' (Critical stress intensity factor). Relationships between these quantities are empirical, but have been well validated over many years for structural steels in moderate section thickness, permitting Charpy impact data to be used to assess lower-bound fracture toughness values.

Operating at a temperature above the ductile-to-brittle Charpy transition temperature does not automatically guarantee the avoidance of brittle fracture in a structural component. Charpy results cannot be considered to be directly relevant to structural behaviour. The apparent changes in toughness that result from the geometry and strain rate are not quantified at all by the Charpy test, which always uses a standard small (10mm thick) specimen under dynamic loading. Many codes deal with these effects by specifying a difference between the Charpy test temperature and the design temperature of the structure. To extrapolate the relationship between Charpy values and fracture toughness to (moderately) stronger steels, some codes increased the required minimum-absorbed-energy value according to the ratio of the yield stress values (on the basis that the design stress limit is a fixed fraction of the SMYS value). This was soon simplified to the '$R_e/10$' criterion.

Modern steel-making techniques and processes have extended the range of tough weldable steels upwards in terms of strength, and have produced significant improvement in the toughness levels of steels, particularly at the lower end of the high-strength range. This raises doubts about the applicability, relevance and conservatism of recommendations and limits in those codes of practice written for the older steels and validated by data from them. It also raises questions about the extrapolation of limits to deal with higher strength steels.

Modern high strength steels often use grain size refinement to increase the yield strength and have grain-growth controllers to maintain the properties as much as possible in the heat affected zones of welds. Usually, the toughness of these steels is also excellent, since fine grain size enhances this property too.

For ultra-high strength steels, extra alloying additions are needed to give the strength and these are likely to raise the ductile-to-brittle transition temperatures. Weld metals for use with high strength steel are required to show comparable levels of strength and therefore have to develop strong tough structures on solidification, which also requires alloying additions. Very high toughness, therefore, may be less readily achieved in these two cases.

While the strength of steel has increased, it is not clear whether the same temperature offset (to take account of different structural thickness) between the design temperature and the Charpy impact test temperature is equally applicable to the higher strength steels. The same doubts apply to the nominal correlation between fracture toughness and a required energy level in Charpy tests (27J, '$R_e/10$' etc). For 690MPa steel, IACS has agreed on a requirement of 69J at –60°C, as already detailed in section 4.

Where the same relationships between impact testing and fracture toughness are extended to thick-sectioned material and to high strength steel, specifications should be viewed with caution until it has been demonstrated that adequate factors of reserve are incorporated for the new conditions. A new European structural integrity assessment procedure is incorporating a statistical approach so that the effect of these uncertainties on the reliability of the assessment can be quantified, but it is not specifically aimed at high strength steel.

Direct testing for fracture toughness may be preferable to Charpy correlation, particularly since it can reduce some of the uncertainties related to the effect of material thickness.

Because of the changes in both the position and steepness of the ductile-to-brittle transition curve for the fracture toughness, it may be unwise to rely on a single point - especially a Charpy point - to characterise the transition. It is more prudent to investigate where the fracture transition temperature actually lies in relation to the design temperature.

Fracture testing can be performed on the full thickness of the structural material, and hence improve confidence in the relevance of the numbers used as input to any assessment. Some uncertainty in applying the results will still arise where the geometry is complex and where the residual stresses are difficult to define.

Designing from K_{IC} values is an option, but the recommendations are likely to be conservative for thin-sectioned high-toughness steel. Although thick sections are capable of developing plane strain conditions making it difficult for plasticity to occur at the crack tip, especially for higher strength steel, it does not mean that plane strain conditions always apply in thick sections. A surface crack that is not very deep may not experience constraint of yielding, hence its apparent toughness may be significantly above K_{IC}.

Environmental influences must be taken into account. K_{ISCC} for example may be appreciably lower than the K_{IC} value.

Because of the toughness improvements that have accompanied increases in steel strength, it is possible that the limitations to the use of high strength steel will shift from fracture behaviour, towards buckling and crack extension from mechanisms such as fatigue. This means that it may be necessary to sacrifice the full capabilities of increasing the design stress in a structure in line with the increase in SMYS. Instead, the applied stress could be increased in line with the more moderate improvements in fatigue resistance etc, which may still give appreciable benefits in weight-saving and reduced fabrication costs, as well as enhanced resistance to fracture.

REFERENCES

6.01 *Stacey A, Sharp J V and King R N,* 'High strength steels used in offshore installations', Proc. 15th Intnl. Conf. OMAE, Florence, June 1996, ed. M.M. Salama et al, publ. ASME, Vol. III, pp 417-433

6.02 BS 7448: 1991 'Fracture mechanics toughness tests' Part 1: 'Method for determination of K_{IC}, critical CTOD and critical J values of metallic materials', British Standards Institution, London

6.03 BS 7448: 1997 'Fracture mechanics toughness tests' Part 2: 'Method for determination of K_{IC}, critical CTOD and critical J values of welds in metallic materials', British Standards Institution, London

6.04 *Kaiser H-J, Kern A, Schriever U and Wegmann H,* 'Fracture toughness of modern high-strength steel plates with minimum yield strength up to 690MPa'. In: Proceeding of International Symposium 'Safety in application of high strength steel', Trondheim, Norway, 1-2 July 1997, publ. Statoil, Norway

6.05 *Bennett W T, Cadiou L and Coudreuse L,* 'Steels for jack-up legs'. In: 'Recent Developments in Jack-up Platforms', ed. Boswell L F and D'Mello C, Blackwell Scientific Publications, Oxford, 1992, Ch 12, pp283-312

6.06 Offshore Technology Report OTN 96 127, 'Toughness acceptance criteria for rack materials in jack-ups', Health and Safety Executive, July 1996 [Restricted]

6.07 Det norske Veritas Report 95-3056, 'Toughness acceptance criteria for rack materials and weldments in jack-ups', 1995, publ. Det norske Veritas, Oslo, Norway

6.08 *Sumpter J D G,* 'Fracture safety of high strength steels'. In: Proceeding of International Symposium 'Safety in application of high strength steel', Trondheim, Norway, 1-2 July 1997, publ. Statoil, Norway

6.09 IACS Requirements, 1994 Rev 1 Section W16, 'High Strength Quenched and Tempered Steels for Welded Structures'

6.10 BS 7910:1999 'Guide on methods for assessing the acceptability of flaws in fusion welded structures', British Standards Institution, London

6.11 BS EN ISO 12737:1999, 'Metallic materials: Determination of plane-strain fracture toughness', British Standards Institution, London

6.12 *Cotton H J,* 'Application of Fracture Analysis in the Design of Fixed Offshore Structures'. In: 'Materials Technology symposium', Proc. 19th Intnl. Conf. OMAE, New Orleans, USA, February 2000, publ. ASME

6.13 *Wallin K and Nevasmaa P,* 'Structural Integrity Assessment Procedures for European Industry (SINTAP) - Sub-task 3.2 report : Methodology for the treatment of fracture toughness data – procedure and validation', Report VAL A: SINTAP/VTT/7, VTT Manufacturing Technology, Espoo, Finland, 1998

6.14 *Bannister A C,* 'Structural Integrity Assessment Procedures for European Industry (SINTAP) - Sub-task 3.2 report : Determination of fracture toughness from Charpy impact energy – procedure and validation', Report SINTAP/BS/15, British Steel plc, Rotherham, UK, 1997

6.15 *Nevasmaa P, Bannister A C and Wallin K,* 'Fracture toughness estimation methodology in the 'SINTAP' procedure'. In: Proc. 17th Intnl. Conf. OMAE, Lisbon, June 1998, Paper OMAE98-2053, publ. ASME

6.16 *Webster S E and Bannister A C,* 'Structural Integrity Procedure for Europe (SINTAP): The Complete Story'. In: Proc. 18th Intnl. Conf. OMAE, St John's Newfoundland, July 1999, Paper OMAE99-MAT2040, publ. ASME

6.17 *Bannister A C and Stacey A,* 'Literature Review of the Fracture Properties of Grade A Ship Plate'. In: Proc. 18th Intnl. Conf. OMAE, St John's Newfoundland, July 1999, Paper OMAE99-MAT2071, publ. ASME

6.18 *Minami F, Jida M, Takahara W, Kondo N and Arimochi K,* 'Fracture Mechanics Analysis of Charpy Test Results based on the Weibull Stress Critereion'. In: Proc. 20th Intnl. Conf. OMAE, Rio de Janeiro, June 2001, Paper OMAE 2001/MAT-3130, publ. ASME

6.19 *Hancock P, Spurrier J and Chubb J P,* 'The problems of weld metal or heat affected zone toughness in offshore structural steels', Proc. 15th Intnl. Conf. OMAE, Florence, June 1996, ed. M.M.Salama et al, publ. ASME, Vol.III, pp333-339

Figure 6.1 – Charpy V-notch transition temperatures for some modern Thyssen steels [6.04]

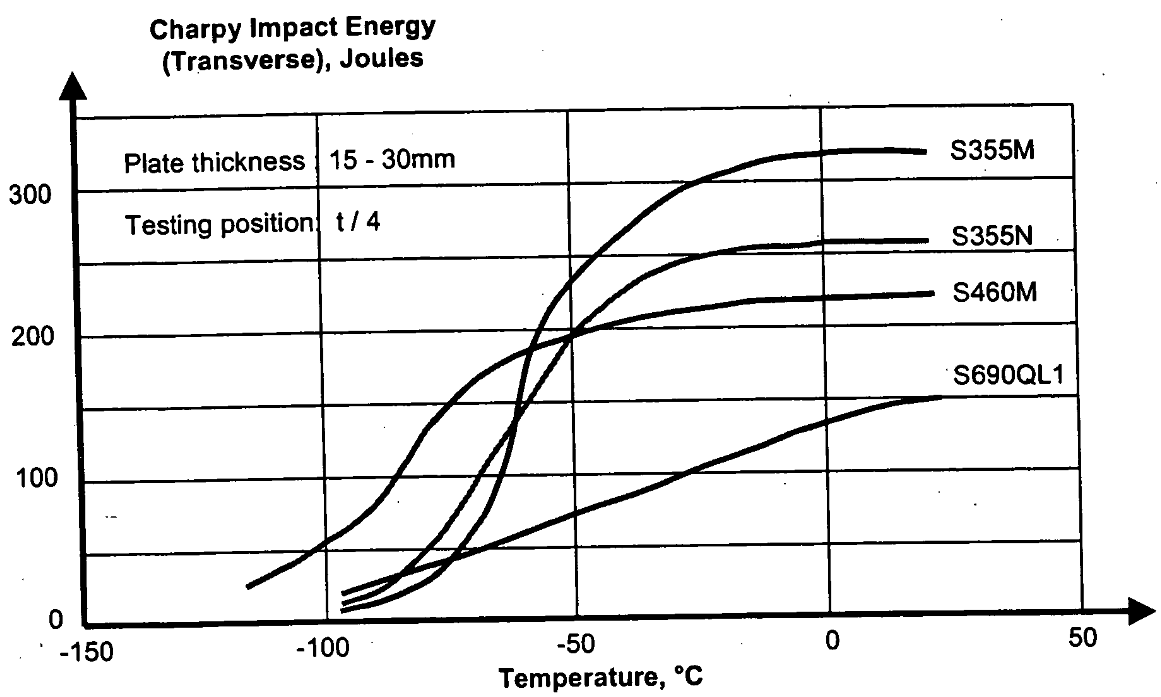
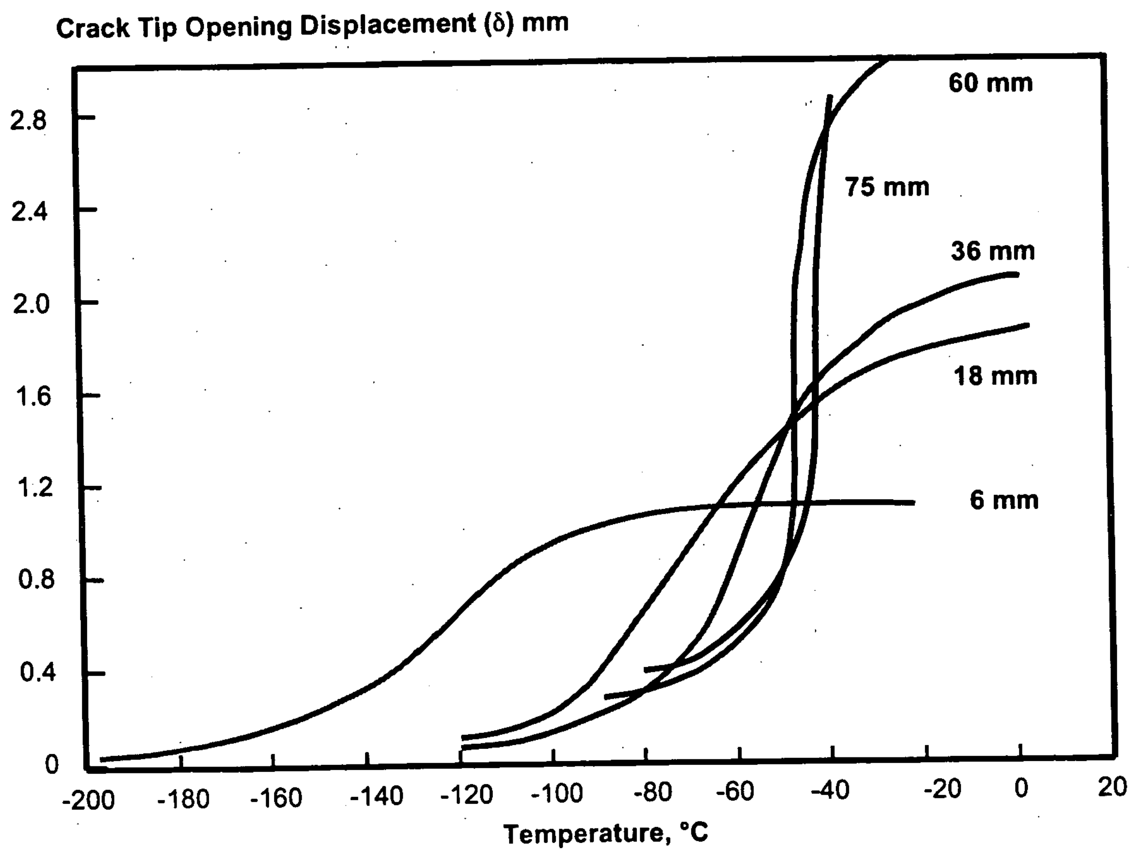


Figure 6.2 – Crack tip opening displacement transition temperatures for geometrically similar samples from the same plate of steel, with crack tips at constant depth in the plate [6.19]



7. FATIGUE IN HIGH STRENGTH STEELS

7.1 INTRODUCTION

Design to resist fatigue is recognised as one of the main requirements for offshore structures, particularly for welded tubular joints in seawater and subject to high stress concentrations. The procedure is well established for medium strength steels, with appropriate S-N curves published in design codes. However, the effect of seawater on the fatigue performance is considered to be more detrimental for high strength steels because of their greater susceptibility to hydrogen cracking. This susceptibility is known to increase with increasing yield strength and more negative cathodic protection (CP) potentials. Hydrogen generation from CP can enhance crack growth rates at the crack tip, leading to overall shorter fatigue lives. Hence, there is a need to understand the fatigue process in welded high strength steels. However, there is a very limited amount of data for the higher strength steels which makes it difficult to provide design information (see section 13).

Corrosion fatigue is a major cause of failure in marine structures with most of the fatigue life being taken up in fatigue crack propagation. The corrosion fatigue behaviour of welded joint constructions from structural steel conforming to BS4360:50D has been the subject of many studies over the years [7.01]. As a result, the understanding of joint geometry, seawater environment and CP has reached a level where confident predictions on fatigue resistance behaviour can be made for this type of steel/structure.

In addition to S-N type data there is a need to estimate the remaining lives of components containing cracks, using fracture mechanics. There is an increasing amount of data on crack growth rates (da/dN) versus stress intensity factor range (ΔK), for all types of steel, both in-air and seawater, which has enabled the relevant design code (BS 7910) [7.02] to cover steels with yield strengths up to 600MPa. However, for even higher strength steels, the data, particularly in seawater under CP, remain limited and special approaches are recommended, usually based on test data for the steel in question.

7.2 FATIGUE CRACK PROPAGATION

7.2.1 Effect of steel strength on fatigue crack growth rate

King [7.03, 7.04] carried out a review of literature FCGR for structural and engineering steels with yield strengths up to 1000MPa for tests in-air and seawater environments (free corrosion and CP levels of -800 and -1050/1100mV(Ag/AgCl)). The data were split into two strength ranges of ≤ 450 and >450MPa and R ratios of ≥ 0.5 and 0-0.1 and were also limited to cycling frequencies of <1Hz. For each data set the intercept (C) and gradient (m) values from the mean linear regression line fitted to the Paris Law were calculated. The standard deviation was calculated for each curve and then used to calculate C values corresponding to the ± 2 standard deviation ($\pm$2SD) curves. Stage 1 and Stage 2 growth were allowed for.

The data showed that there was no obvious effect of yield strength on fatigue crack growth rates (FCGR) even for seawater with CP. This was an unexpected finding as the higher strength steels are commonly expected to have a greater susceptibility to hydrogen effects than the lower strength steels. The review only considered parent plate material and therefore it is suggested that caution should be taken when applying the data to high strength steel heat affected zones (HAZ) and weld metals. In addition, the effects of sulphides/SRB and hydrogen precharging were not included.

It was decided to use the $\pm$2SD lines produced by this review as the baseline for comparison of FCGRs on the da/dN versus ΔK plots in this review. The Paris Law constants used and the crossover points for stage 1 and 2 growth are listed in Table 7.1. Based on this review BS7910 has provided data for steels with yield strengths up to 600MPa. Although the review covered even higher strength steels, it was felt that the limited data available did not justify increasing the limit in BS7910 above 600MPa.

7.2.2 Parent Materials

The King [7.03, 7.04] review covered parent material and found that overall the behaviour of high strength steels was similar to the medium strength steels under the conditions reviewed. It was surprising that even with cathodic overprotection the behaviour was still similar. Several fractographic studies have confirmed the change in crack propagation mode, from a ductile mechanism with secondary cracking to quasi-cleavage fracture, at high negative CP potentials [7.01]. Healy [7.01] reviewed the FCGR of high strength steels and found that the performance of the high strength steel group was comparable to, if not slightly improved over, that of the structural grade BS4360:50D steels. For the steels examined, no discernible effect of manufacturing process (i.e. between Q&T and TMCP) was seen on the resultant FCGR.

7.2.3 HAZ

Very limited fatigue data have been generated for the HAZ of high strength steel weldments. Representative HAZ FCGR data for steels in the strength range 500-600MPa are presented in Figure 7.1. The available data show that the fatigue performance of the HAZ is similar to that observed for both the structural grade steel and the high strength parent plate data. Extensive metallographic and fractographic studies have shown that high strength steels with yield strengths in the range 500-600MPa are not susceptible to excessive hardening in the welded condition (HAZ hardness generally <350Hv [7.05]). Additionally, the fatigue crack failure mechanisms in the HAZ are similar to those observed in the parent plate, displaying a similar response to environmental test conditions. It would appear that welding under controlled conditions does not significantly affect FCGR in these steels and the data suggest that a preliminary screening of the corrosion fatigue behaviour on the parent plate may be used to assess the suitability of the steel when welded.

7.2.4 Weld Metals

There is very little fatigue data available for high strength steel weld metals. Work at Cranfield University [7.05, 7.06] has produced data for weld metals produced by SAW and FCAW in-air and in seawater with CP. The data are shown in Figures 7.2 to 7.6 and shows that the behaviour of the weld metals is comparable to parent materials. Tests were also carried out on a 450MPa steel (not shown in the Figures) and these were comparable to the higher strength weld metals. No relationship between yield stress and fatigue performance was found for the weld metals tested. Again it would appear that welding under controlled conditions does not significantly affect FCGR in these steels and the data suggest that a preliminary screening of the corrosion fatigue behaviour on the parent plate may be used to assess the suitability of the steel when welded.

7.2.5 Fatigue Thresholds

Limited information has been reported on threshold stress intensity values for high strength steels [7.01]. For low R ratios an apparent increase in threshold value occurs with increased levels of CP compared with the in-air behaviour. Many studies have demonstrated that this behaviour is due to crack wedging effects reducing the effective stress intensity range. Under conditions of high load ratio, this mechanism is diminished and threshold values are similar to in-air levels. King [7.03, 7.04] performed a review of fatigue threshold values for carbon and carbon manganese steels and the data are shown in Figure 7.7. It was suggested that the existing recommendations for steels of yield strength <600MPa in PD6493:1991 should be retained. For high strength steels there is insufficient data to make additional recommendations.

7.3 EFFECT OF SRB AND SULPHIDES

Robinson [7.07] reviewed the literature data for sulphide/SRB influenced corrosion fatigue of high strength offshore steels with yield strengths in the range 350 to 1010MPa that included parent material and welds. The data were grouped on the basis of sulphide level, CP potential and R ratio and, for each group, the stage 2 mean line was calculated (steels were also sub-grouped by yield strength). These mean lines have been added to Figures 7.3 to 7.6.

The effect of sulphides/SRB on fatigue threshold values is unclear due to the lack of data. Some workers [7.08, 7.09] have looked at near threshold values and found that the combination of CP and

sulphides/SRB gave threshold values comparable to or greater than those for in-air. It is thought that the deleterious effect of increased hydrogen charging is balanced by a scale induced crack closure effect reducing the effective ΔK. Whilst these results are promising it should be noted that the above tests were carried out under constant amplitude loading and that variable amplitude loading may produce a different result.

It is clear from Figures 7.2 to 7.6 that the hydrogen uptake that results from the combined effects of sulphides and CP has caused increased FCGR in the steels tested. The data suggest that saturated H_2S is the worst case but that much lower concentrations can also give significant increases in FCGR. The effect of yield strength on FCGR in sulphide containing environments is not clear due to the lack of data. Overall accelerated FCGR occur in both medium and high strength steels when exposed to sulphide and higher rates are found with increasing sulphide concentration and/or more negative CP potentials.

7.4 SN DATA

The amount of information available for steels with yield strengths <500MPa is considerable but becomes more limited for steels with yield strengths $\geq$ 500MPa. Data for in-air tests and seawater with applied CP are given in Figures 7.8 - 7.10. The data have been thickness corrected to 16mm using a thickness exponent of 0.3.

Most of the available data are for in-air testing and include constant and variable amplitude testing of tubular joints and plate specimens constructed from a range of steels and strength levels. In Figure 7.8, data that falls below the 'T' curve (mainly 810 – 840MPa strength level) originate from plate testing for which the appropriate curve is Class F2. It can be seen that all the plate data lies on or above the F2 curve. The data available for tests in seawater with CP are very limited [7.10]. In Figures 7.9 and 7.10 data falling below the air 'T' curve are plate tests and therefore Class F2. At present the data suggest that the fatigue performance of the higher strength steels is generally good but more data are required for steels with applied CP.

7.5 FATIGUE IMPROVEMENT TECHNIQUES

Weld fatigue improvement methods can be divided into two main groups [7.11]. Firstly weld geometry modification which removes toe defects and/or reduces the stress concentration, and secondly, residual stress methods which introduce compressive stress in the area where cracks are likely to initiate. The methods are summarised in Figure 7.11. Many tests have been carried out to establish the gain in fatigue life as a result of using these methods. However most of these have been on grade BS4360:50D type steels. As a result of these data, current codes and standards allow benefits for some of these methods. One of the restrictions in providing benefits in codes is the quality control aspect when using the technique in the field. Table 7.2 lists the improvement techniques which are allowed in current offshore codes for medium strength steels. In most cases there are requirements to be met to achieve the benefit (e.g. quality control, inspection, adequate CP).

For higher strength steels (>500MPa yield strength) there is a very limited amount of data. Work by Bell et al [7.12] was carried out on steels with yield strengths of both 350 and 550MPa using T type joints with longitudinal fillet welds. The thickness of the base plate was either 18 or 26mm. All the tests were in-air, with fatigue lives up to 10^8 cycles. Specimens that had been hammer peened showed a significant gain in life. In the case of the 550MPa steel the average gain was 175%. However cracking in the hammer peened higher strength joints initiated at the root of the weld rather than the weld toe (the location of initiation for the as-welded joints). Table 7.3 summarises other data obtained for a number of improvement techniques for high strength steels. All the data are in-air and show a range of improvements but unfortunately no tests were carried out in seawater.

At present there are insufficient data to demonstrate the benefits of improvement techniques for high strength steels, despite the possible benefits. It would be necessary therefore for a potential user to

demonstrate any benefits using a test programme which represented the conditions in which the steel would be used in service.

7.6 SUMMARY

The fatigue data available for parent and welded high strength steels indicate that the general performance of the high strength steels is as good as the medium strength steels. The only condition where poor performance was found was with H_2S but medium strength steels also show similar poor performance. Weld improvement techniques show promise but require data for typical offshore conditions. In all areas more data are required before confident predictions of the fatigue performance of high strength steels can be made. At the present time producing test data for candidate high strength steels still appears to be the best approach.

REFERENCES

7.01 *Healy J, Billingham J, Stacey A, Simpson R and Patel R* – 'Review of the corrosion fatigue performance of medium to high strength structural steels' Proceedings 15th International Conference Offshore Mechanics and Arctic Engineering, Florence, Italy, 16-20 June 1996, ASME, pp451-461

7.02 British Standard' Guidance on methods for assessing the acceptability of flaws in welded structures, BS 7910, 1991

7.03 *King R N* – 'A review of fatigue crack growth rates in air and seawater' Offshore Technology Report OTH96 511, HSE (1996)

7.04 *King R N, Stacey A and Sharp J V* – 'A review of fatigue crack growth rates for offshore steels in air and seawater environments' Proceedings 15th International Conference Offshore Mechanics and Arctic Engineering, Florence, Italy, 16-20 June 1996, ASME, pp341-348

7.05 *Kilgallon P, Healy J and Billingham J* – The corrosion fatigue behaviour of high strength steels and associated high strength weld metals. Offshore Technology Report, OTO97 065, HSE (1997)

7.06 *Billingham J, Blackman S and Norrish J* – Further assessment of high strength steel weld metals for use in offshore engineering applications. Offshore Technology Report OTO98 091, HSE (1998)

7.07 *Robinson M J and Kilgallon P J* – A review of the effects of sulphate reducing bacteria in the marine environment on the corrosion fatigue and HE of high strength steels' Health and Safety Executive, Offshore Technology Report OTH 98 555, HMSO

7.08 *Ferguson W G, Zhang Y, Stevens F J and Assefpour-Dezfuly M* – 'The effects of an anaerobic environment on corrosion fatigue' Proceedings of Chemical '90, 18th Australasian Chemical Engineering Conference, 27-30 August 1990, Auckland, New Zealand, pp286-295

7.09 *Rudd W J and Booth G S* – 'Near threshold crack growth in structural steels' British Steel Report FR 5330-8 942 May 1995

7.10 *N Tantbirojn, L S Etube, W D Dover, P J Kilgallon, T Roberts and J Spurrier* – 'Variable Amplitude Corrosion Fatigue of Jack-up Steels (VACF-T: Thick Plate Specimens), Final Report, UCL NDE Centre, July 2000

7.11 *Kirkhope K J, Bell R, Caron L, Basu R I and Ma K-T* - Weld detail fatigue life improvement techniques. Part 1: review. Marine Structures 12 (6), pp447-474 (1999)

7.12 *Bell R, Militaru D V, and Braid J E M,* 'The Fatigue Life Improvement of High Strength Steel Welded Joints using hammer peening techniques', OMAE conference, 1995, Vol. III, ASME

7.13 *Bignonnet A, Picouet L, Lieurade H P and Castex L* – 'The application of shot peening to improve the fatigue life of welded steel structures' Steel in Marine Structures. Developments in Marine Technology, 3. Proceedings, 3rd International ECSC Offshore Conference (SIMS '87), Delft, 15-18 June 1987. Ed: C.Noordhoek, J.de Back. Publ: 1000 AE Amsterdam, Netherlands; Elsevier Science Publishers BV; 1987. ISBN 0-444-42805-4. Paper SIMS TS 33. pp669-678

7.14 Dept. of Energy, 'Offshore Installations - Guidance on Design, Construction and certification', HMSO, London 1990

7.15 NORSOK, 'Design of Steel Structures' N-004, 1998

7.16 ISO ' Petroleum & Natural Gas Industries, 'Offshore Structures - Part 2'., ISO 13819-2, to be published

7.17 *Lopez Martinez L, Blom A F, Trogen H and Dahle T* – 'Fatigue behaviour of steels with strength levels between 350 and 900 MPa - influence of post weld treatment under spectrum loading' Welded High-Strength Steel Structures. Proceedings, First North European Engineering and Science Conference (NESCO 1), Stockholm, Sweden, 8-9 Oct.1997, Ed: AF Blom, Engineering Materials Advisory Services Ltd pp361-376 (1997)

7.18 *Agerskov H* - Fatigue in steel structures under random loading. Journal of Constructional Steel Research, 53 (3) March pp 283-305 (2000)

7.19 *Agerskov H, Petersen R I and Lopez Martinez L* – 'Fatigue in high-strength steel offshore tubular joints' Tubular Structures VI, Proceedings of 6th International Symposium on Tubular Structures, Melbourne, Australia 14-16th December 1994, pp527-534

Table 7.1 – Data used to construct lines on Paris fatigue plots

Environment (v. Ag/AgCl)	R	m		C						Stage 1/2 changeover ΔK $Nmm^{-3/2}$		
				Mean		+2SD		-2SD				
		Stage 1	Stage 2	Stage 1	Stage 2	Stage 1	Stage 2	Stage 1	Stage 2	Mean	+2SD	-2SD
Air	0 – 0.1	8.16	2.88	1.21×10^{-26}	3.98×10^{-13}	4.37×10^{-26}	6.77×10^{-13}	3.35×10^{-27}	2.34×10^{-13}	363.1	314.9	418.8
	≥0.5	5.10	2.88	4.80×10^{-18}	5.86×10^{-13}	2.10×10^{-17}	1.29×10^{-12}	1.10×10^{-18}	2.66×10^{-13}	195.6	143.5	266.5
Free corrosion	0 – 0.1	3.42	1.30	3.00×10^{-14}	1.27×10^{-7}	8.55×10^{-14}	1.93×10^{-7}	1.05×10^{-14}	8.36×10^{-8}	13336.0	993.1	1797.3
	≥0.5	3.42	1.11	5.37×10^{-14}	5.67×10^{-7}	1.72×10^{-13}	7.48×10^{-7}	1.68×10^{-14}	4.30×10^{-7}	1097.8	747.7	1611.7
-850mV	0 – 0.1	8.16	2.67	1.21×10^{-26}	5.16×10^{-12}	4.37×10^{-26}	1.32×10^{-11}	3.35×10^{-27}	2.02×10^{-12}	462.2	434.0	492.2
	≥0.5	5.10	2.67	4.80×10^{-18}	6.00×10^{-12}	2.10×10^{-17}	2.02×10^{-11}	1.10×10^{-18}	1.78×10^{-12}	322.9	289.9	359.6
-1050mV	0 – 0.1	8.16	1.40	1.21×10^{-26}	5.51×10^{-8}	4.37×10^{-26}	9.24×10^{-8}	3.35×10^{-27}	3.29×10^{-8}	575.6	513.8	644.8
	≥0.5	5.10	1.40	4.80×10^{-18}	5.25×10^{-8}	2.10×10^{-17}	1.02×10^{-7}	1.10×10^{-18}	2.70×10^{-8}	516.7	414.9	643.4

Table 7.2 – Increase on life allowed for weld improvement techniques in codes

Technique	HSE Guidance (7.14)	NORSOK (7.15)	Draft ISO standard (7.16)
Weld profiling	N/A	Increase by factor of 2 on life	N/A
Weld toe grinding	Increase by factor of 2.2 on life	Increase by factor of 2 on life	Increase by factor of 2 on life
TIG dressing	N/A	Increase by factor of 2 on life	N/A
Hammer peening	To be demonstrated by test programme	Increase by factor of 4 on life	Increase by factor of 4 on life

Table 7.3 – Summary of in-air fatigue improvement data (data were obtained using constant (C) and/or variable(V) amplitude loading and improvement was calculated from stress range (S) or cycles (N). * At 2 x 10^6 cycles

Improvement Technique	Steel Type	Yield Strength (MPa)	C/V	S/N	Mean Improvement Factor (%)	Ref
Tig dressed	DOMEX 590	615	CV	N	42	7.17
Tig dressed	WELDOX 700	780	CV	N	73	7.17
Tig dressed	WELDOX 900	900	CV	N	89	7.17
Shot peened	E550	640	C	S*	78	7.13
Hammer peened	HY80	-	-	-	175	7.12
Shot peened	Q&T	730/820	-	S*	70	7.11
Ultrasonic peening	WELDOX 700	780	CV	N	79	7.17
Tig dressed & ultrasonic peening	WELDOX 900	900	CV	N	104	7.17

Figure 7.1 – Summary diagram showing bound curves for HAZ fatigue crack growth rates in 500-600MPa offshore steels

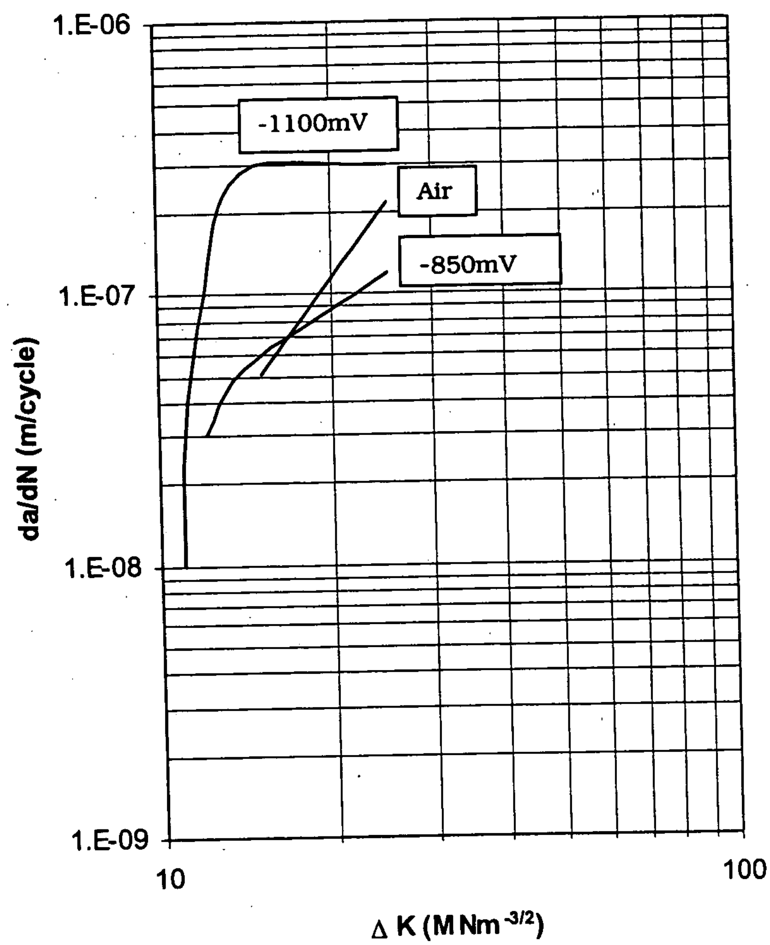


Figure 7.2 – Air fatigue data for RQT501 (open) and WX700 (filled) steels parent plate and weld metals. Black symbols are SAW weld metals and red symbols are FCAW. R 0.5

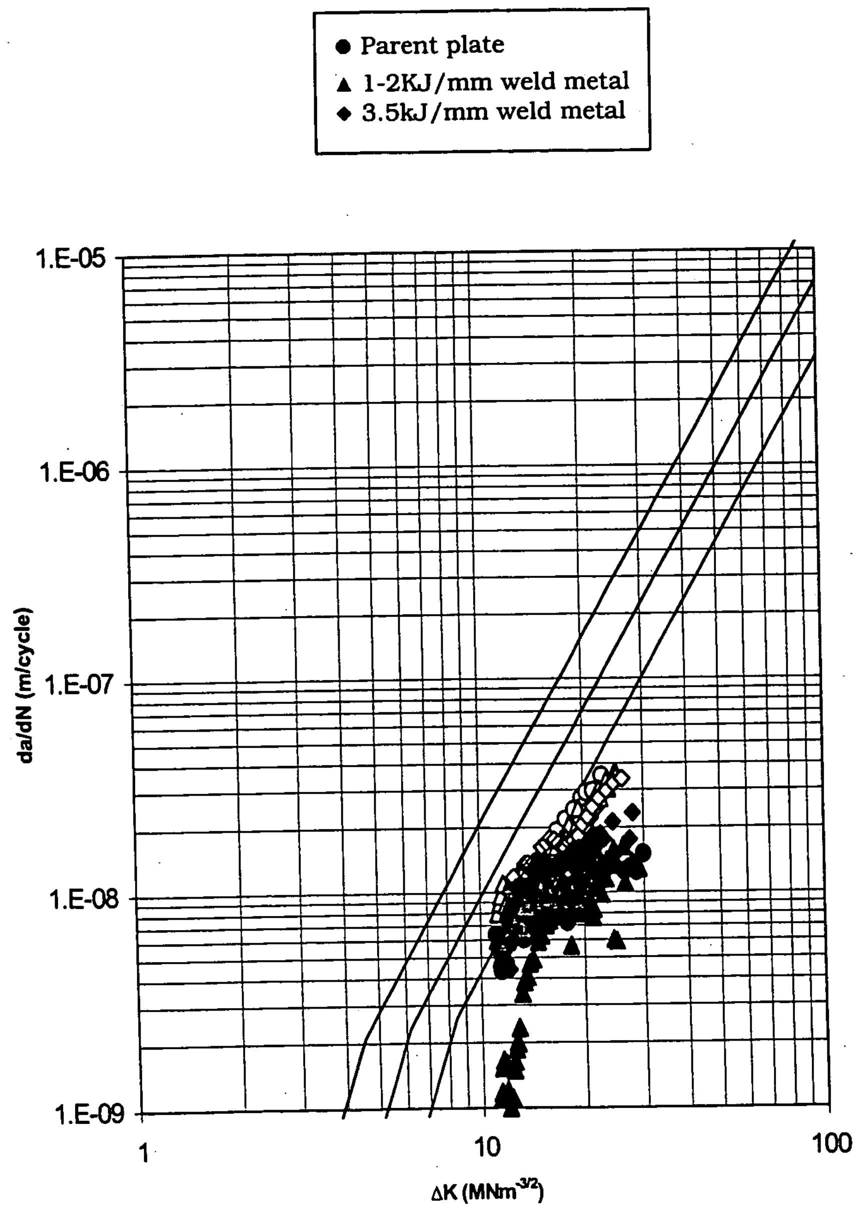


Figure 7.3 – Mean fatigue data for freely corroding steel in seawater containing H_2S at $R \geq 0.5$ [7.07]

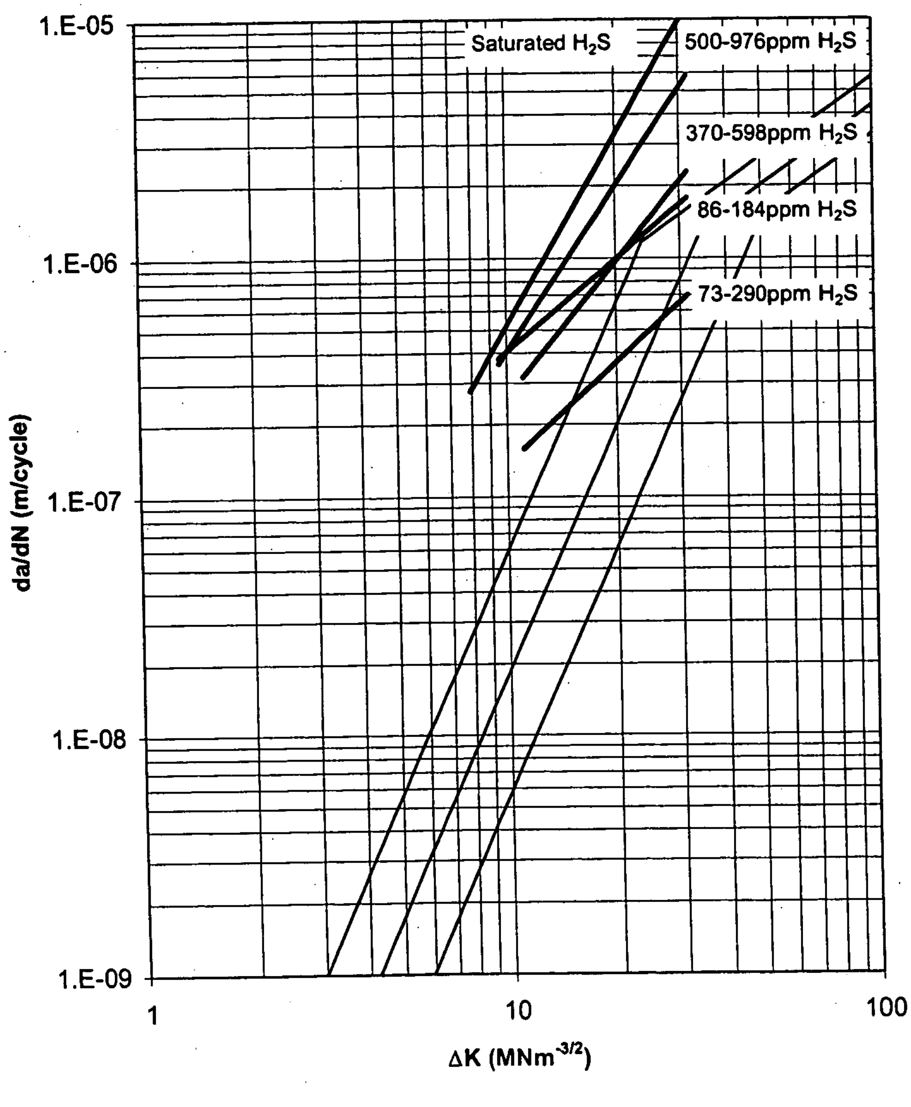


Figure 7.4 – Mean fatigue data for freely corroding and cathodically protected steel in seawater saturated with H_2S at R = 0 – 0.1 [7.07]. Comparison lines are equivalent lines for tests without H_2S

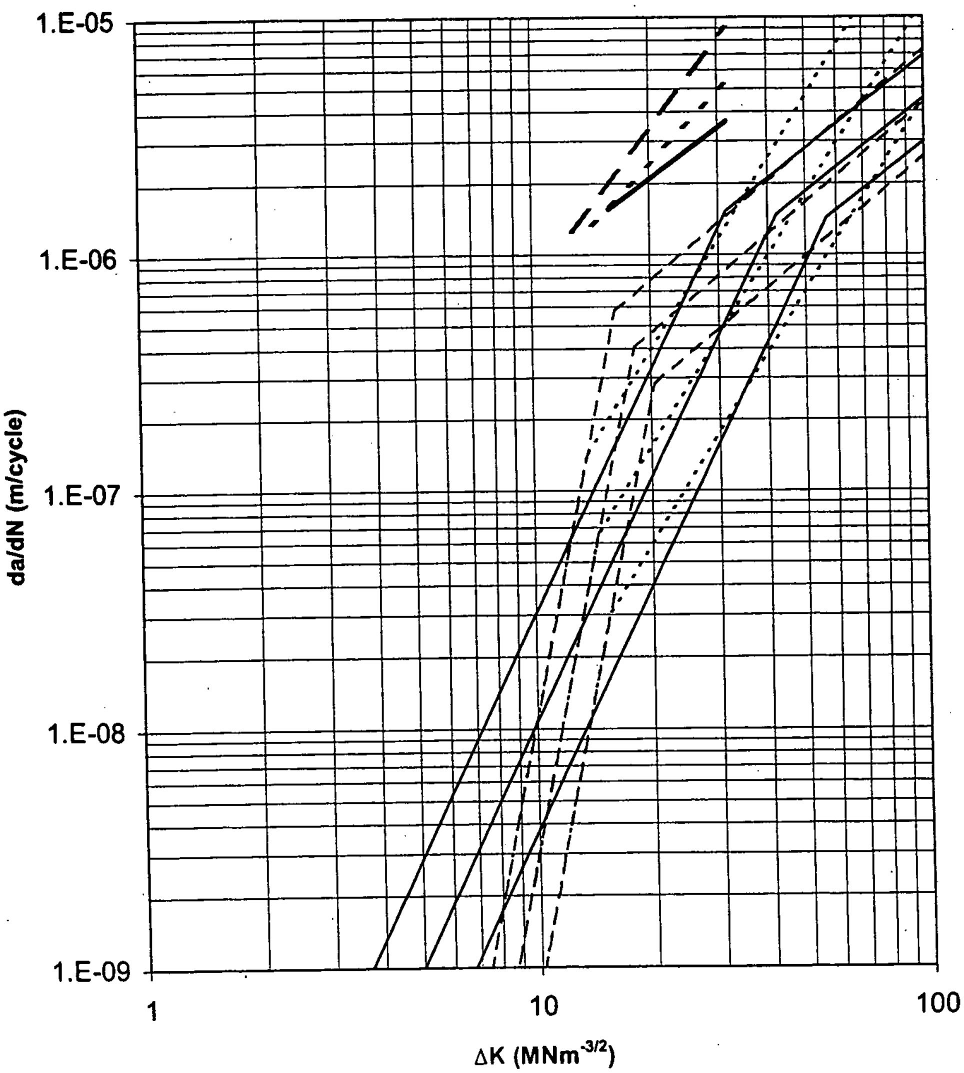


Figure 7.5 – Fatigue data for applied CP of –850mV(Ag/AgCl) for RQT501 (open) and WX700 (filled) steels parent plate and weld metals. Also shown is mean line for fatigue data for tests with saturated H_2S [7.07]

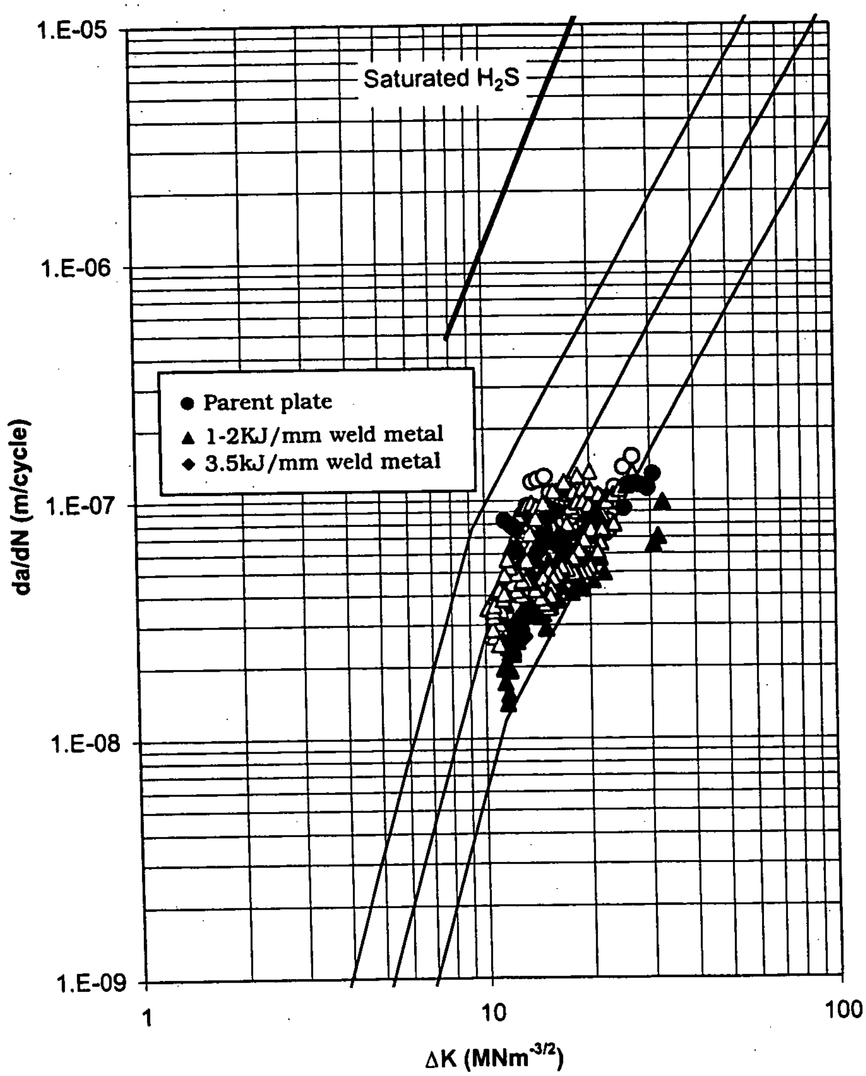


Figure 7.6 – Fatigue data for applied CP of –1100mV (Ag/AgCl) for RQT501 (open) and WX700 (filled) steels parent plate and weld metals. R ≥ 0.5. Also shown is mean line for fatigue data for tests with H_2S [7.07]

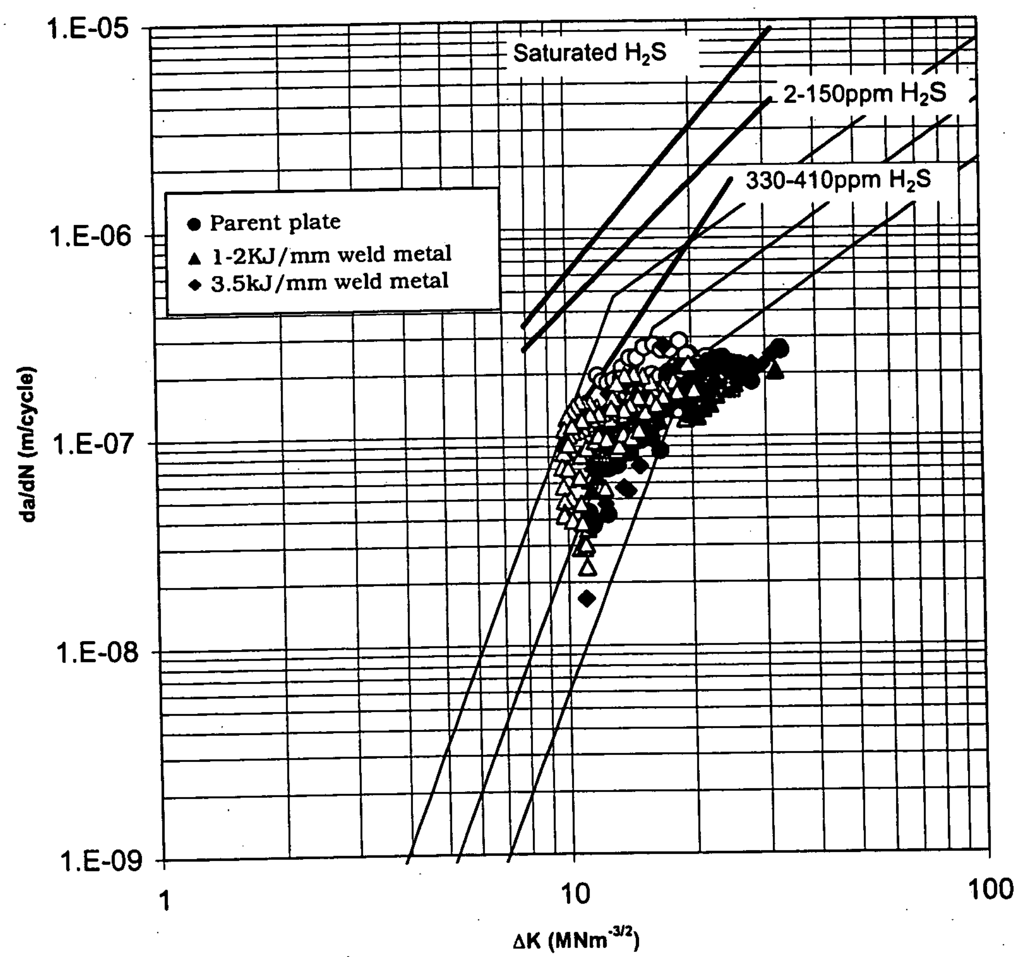


Figure 7.7 – Fatigue thresholds (adapted from 7.03, 7.04)

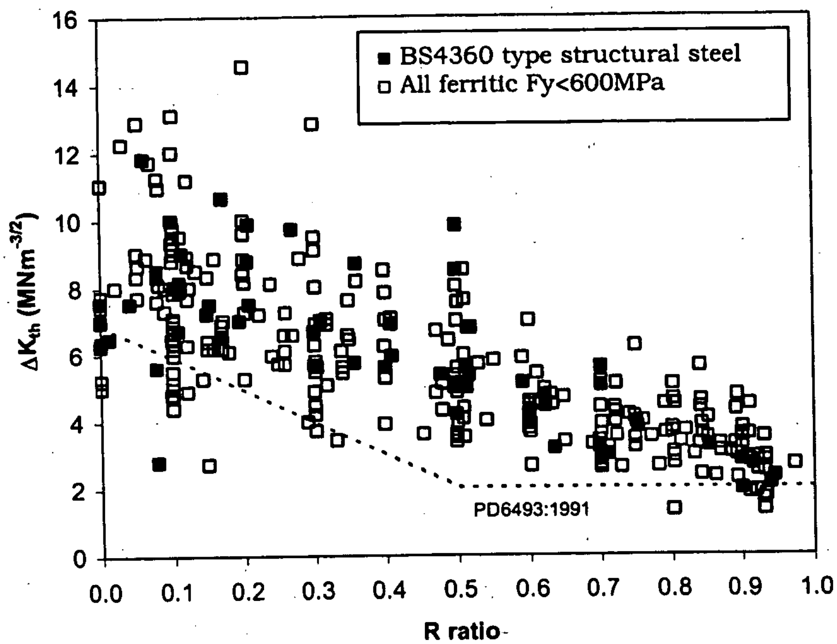


Figure 7.8 – SN data for constant and variable amplitude fatigue tests in-air for parent material and welded steel joints (thickness corrected to 16mm) [7.10, 7.18, 7.19]

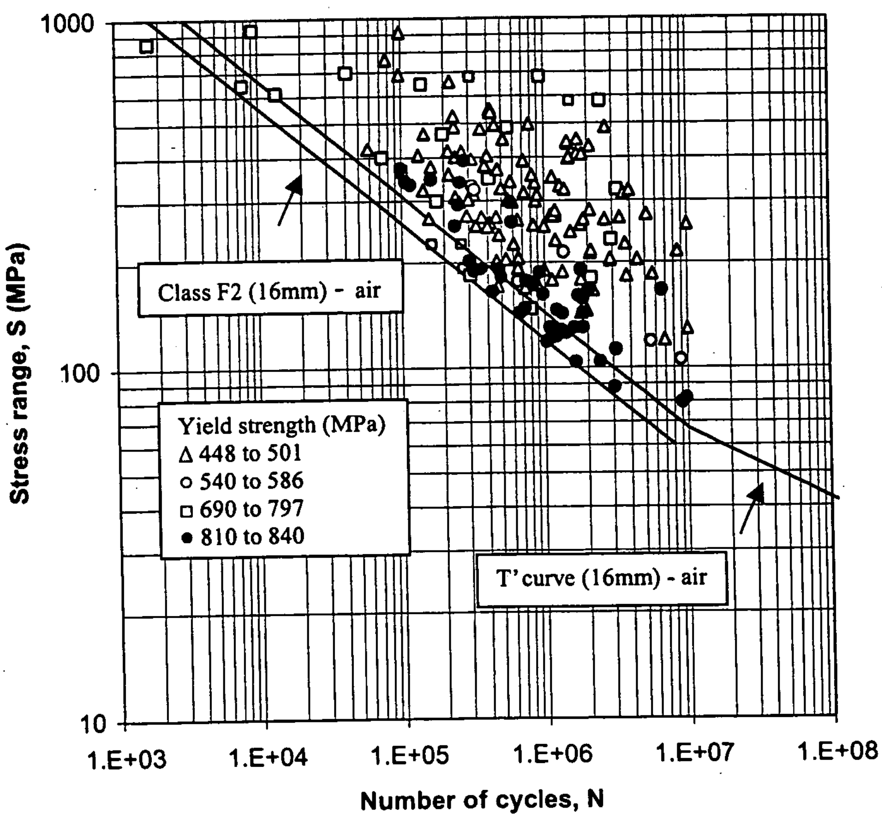


Figure 7.9 – SN data for constant and variable amplitude fatigue tests with applied CP of –800 to –850mV(Ag/AgCl) on welded steel joints

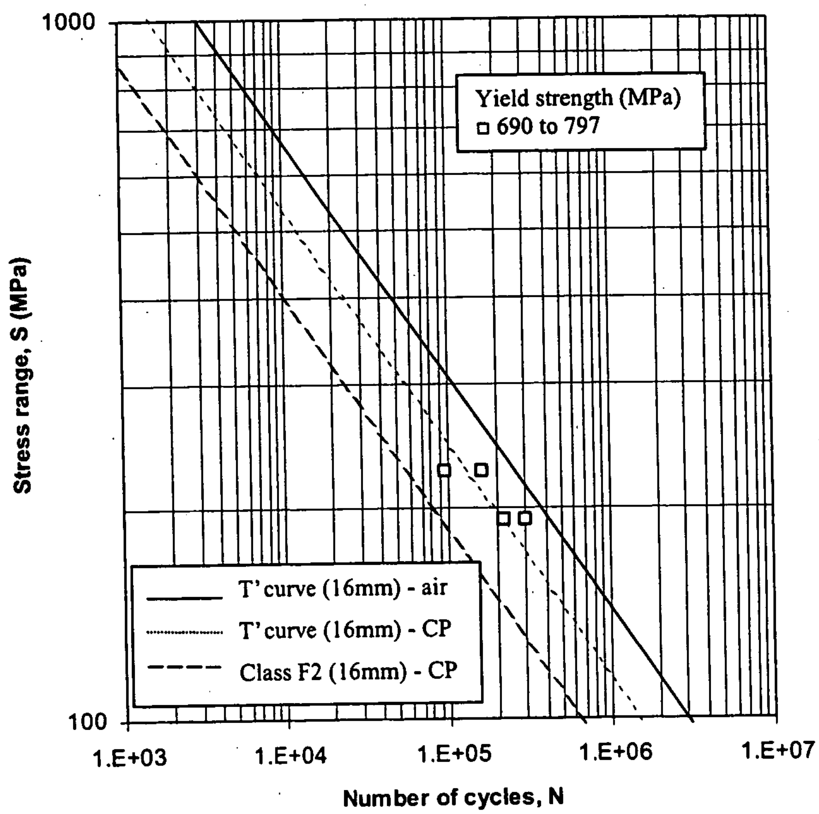


Figure 7.10 – SN data for constant and variable amplitude fatigue tests with applied CP of –1000 to –1050mV(Ag/AgCl) on parent material (shaded) and welded steel joints

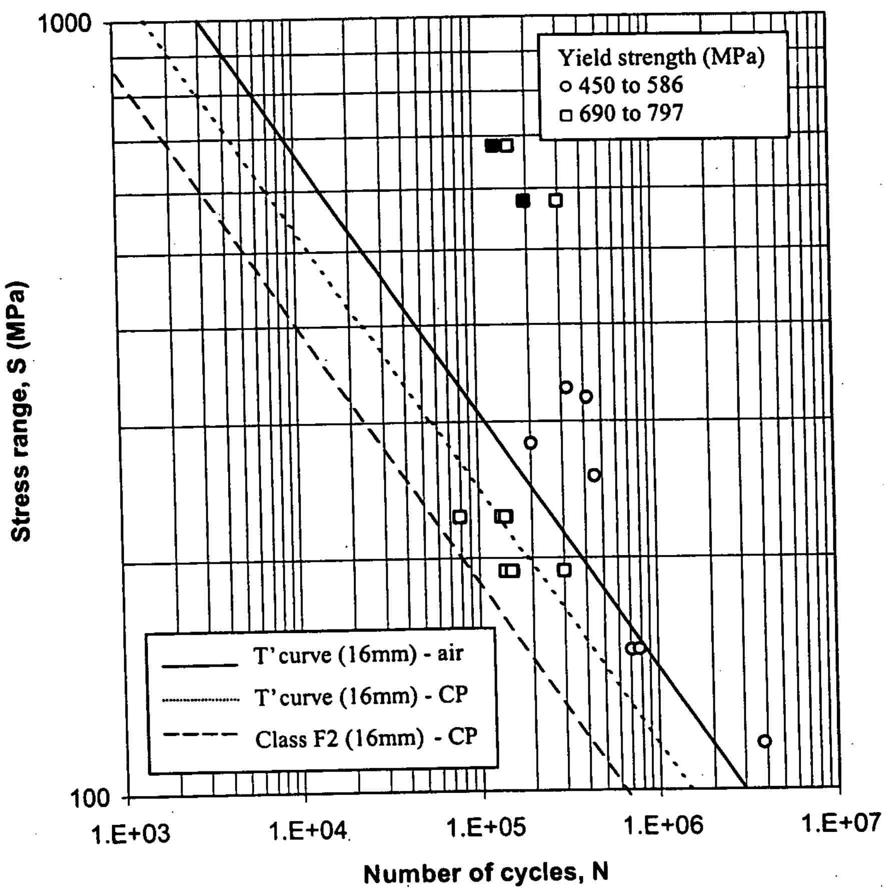
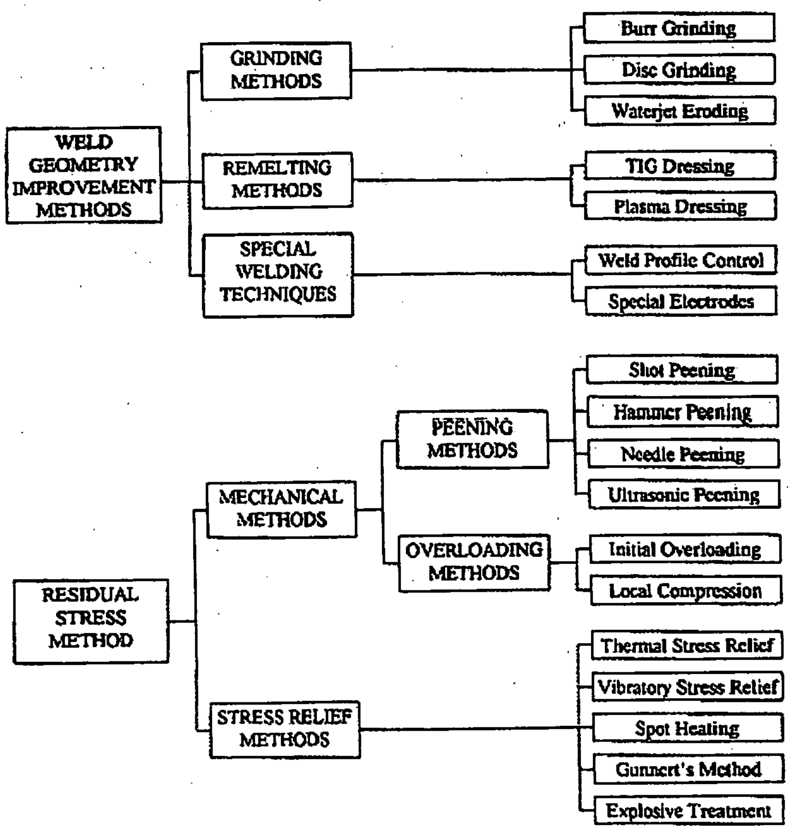


Figure 7.11 – Summary of Weld Fatigue Improvement Methods [7.11]



8. CATHODIC PROTECTION

8.1 INTRODUCTION

There are three systems that can be used to apply cathodic protection (CP), sacrificial anode cathodic protection (SACP), impressed current cathodic protection (ICCP) and hybrid systems of SACP and ICCP. The advantages and disadvantages of these systems are summarised in Table 8.1. CP is used to protect both coated and uncoated steel from corrosion. SACP is the most widely used method for protecting steel structures in the marine environment from corrosion. SACP uses sacrificial anodes (usually aluminium based for structures and zinc based for pipelines) distributed around the structure to ideally give an even distribution of a potential of –850 mV(Ag/AgCl)[2]. Due to the complexity of structures it is possible to have high negative values occurring in the vicinity of anodes more negative than -1000mV(Ag/AgCl) with the risk of hydrogen embrittlement and enhanced fatigue crack propagation, and more positive potentials than -750mV(Ag/AgCl) at remote or shielded locations with the risk of localised corrosion. Clearly, for steels that are susceptible to hydrogen embrittlement it is important to design the CP system to achieve the correct balance between the risks of corrosion and hydrogen damage.

8.2 PROTECTION CRITERIA

There exists a great deal of guidance for CP levels, much of which has been derived empirically. The relevant codes and standards relating to CP of high strength steels are compared in section 4.0. Care must be taken when using this general guidance as it important to realise that the risk of hydrogen cracking depends on the combination of material, loading, CP level and environment. Additionally, the effects of the CP system on other materials connected to the protected steel structure (such as duplex stainless steel) must also be considered.

The CP potential required for full protection (corrosion rate reduced to insignificant level) of steel in seawater is widely considered to be -800mV(Ag/AgCl) [8.01]. Recommended potentials range between -750 and -830mV(Ag/AgCl) [8.02]. Recent work [8.03, 8.04] found that a 700MPa offshore steel was adequately protected (corrosion rate of 0.001mm/year) at potentials in the range -760 to -790mV(Ag/AgCl).

In anaerobic conditions, where active populations of sulphate reducing bacteria (SRB) might be present and producing sulphides, it has been generally recommended, for lower strength constructional steels, that the potential should be lowered further to -900mV(Ag/AgCl) for full protection. Hydrogen will be produced at potentials more negative than -710mV(Ag/AgCl) for North Sea seawater (pH 8.3 & 10°C) [8.05]. It should be noted that the amount of hydrogen produced by CP systems and therefore absorbed by the steel increases as the potential becomes more negative and even small concentrations of sulphide can significantly increase in the amount of hydrogen absorbed by steel [8.06]. These factors have meant that the recommended levels of CP for high strength steels has required special attention due to their greater susceptibility to hydrogen embrittlement, particularly above yield strengths of 700MPa. Recommendations have been made that for steels with yield strengths ≥ 700MPa CP potentials should generally be within the range of -800 to -950mV(Ag/AgCl) [8.01]. For steels with yield strengths >800MPa the potential should not go more negative than -800mV(Ag/AgCl) [8.01]. The ranges described above are illustrated in Figure 8.1.

During the late 1980s, routine surveys of offshore jack-up drilling rigs discovered cracks in the legs and spudcans that were believed to be due to hydrogen embrittlement [8.07]. A subsequent research programme [8.08] included an investigation of the level of CP at which the jack-up steels showed evidence of hydrogen embrittlement. The research employed slow strain rate testing and concluded that to avoid problems the CP potentials should not be more negative than -805mV(Ag/AgCl). A parallel

---

[2] For Ag/AgCl/Cl$^-$ electrode, potential, E (V) = 0.2224 - 0.0591 log a_{Cl^-}, assumed that values given relate to 1M KCl (= 222.4mV) but other concentrations possible, e.g. seawater ~250mV.

study [8.09] demonstrated, using fracture mechanics specimens, that -830mV(Ag/AgCl) was the CP limit for a welded 690MPa yield strength steel in a SRB containing environment (steel loaded to ⅔ yield strength and containing 1mm defect).

A survey of operating protection potentials on offshore platforms using SACP (usually steels with 350MPa yield strengths) [8.01] showed that 65% of the platforms surveyed had potentials more negative than -900mV(Ag/AgCl) and 30% of those surveyed had operating potentials more negative than -1050mV(Ag/AgCl). This survey demonstrates that in practice traditional SACP systems produce CP potentials too negative for high strength steels and thus further steps have to be taken to reduce the risk of hydrogen damage.

8.3 AVOIDING OVERPROTECTION PROBLEMS

The cracking observed in some jack-up designs in the late 1980s highlighted the problems of the combination of high strength steels with overprotection (the effects of SRB were also implicated). It was recognised that CP potential, steel susceptibility and environment had to be considered together. The methods used to reduce the risk of CP related hydrogen embrittlement have been formulated around trying to change one or more of these three factors.

In response to the cracking found in the late 1980s in spudcans, two options were proposed to combat this [8.07]. These were to retain the original anodes and introduce a flushing system or to remove the anodes and add an inhibitor and biocide. Both methods were used initially but eventually the inhibitor/biocide option was adopted.

ICCP or hybrid systems have the advantage that the CP potential can be controlled as long as local potential monitoring is effective. However, these systems have the inherent risk that poor management or system failure could lead to extreme overprotection or no protection. Other disadvantages of ICCP identified by Davey [8.10] are the need for robust reference electrodes, possible interference between zones and maintaining electrical integrity when the jack-up is used for variable depth operation. Davey concluded that these disadvantages made ICCP practical only for jack-ups operating at fixed depth.

There are several ways to reduce the undesirable situation of overprotection by making changes to the SACP system. These methods included the use of dielectric shields, sacrificial coatings, voltage limiting diodes, voltage limiting resistors and low driving voltage anodes.

Dielectric shields are used in ICCP systems to limit potentials close to the anodes. However, for use with sacrificial anodes, the area of shield required is very large and hence not very practical for large offshore structures.

Jack-up legs require corrosion protection for the regions that are not submerged and thus not protected by CP. Coatings provide this protection and can be used for submerged areas together with CP and may offer a practical method of reducing the uptake of hydrogen by steel in the marine environment [8.06]. A suitable coating would reduce the risk of corrosion occurring and lower the current requirements for CP which would give an accompanying reduction in hydrogen uptake. Additionally, the coating could be selected to have antifouling properties. Sacrificial coatings, such as thermal sprayed aluminium, have been utilised and good performance has been reported. Some long term stability and reliability problems have been reported for these coatings and there can also be some additional problems in shielding critical areas from conventional inspection methods, such as magnetic particle inspection. The potentials associated with these coatings are too negative (between -900 and -1000mV(Ag/AgCl)[8.11]) at present.

Anodes can be linked to the structure via a potential limiting diode that ensures that the potential can not become more negative than a set value. These devices were recommended by a Department of

Energy research study [8.08] and have been employed in some offshore production jack-ups, as well as in drilling units. In general, part coating of the structure is required for the diodes to correctly manage the CP levels [8.12]. There is no information on the failure rate of diodes used to control CP but failure of one or several diodes is not expected to cause a severe problem [8.10]. However it has been reported [8.07] that there have been problems in service with a wider range of potentials being found than expected. The field surveys of jack-ups by three certification authorities, reported in reference [8.08], identified that a range of potentials occurred in practice when voltage limited diodes were used (-700 to -900mV(Ag/AgCl)). Problems encountered with CP/diode systems have been explained in some cases by the lack of the correct coating and/or poor modelling [8.12]. The fact that under-protection occurred raises concern over the risk of pitting corrosion and the effect of such potentials on fatigue crack growth rates. This area has received little attention but some results [8.01] do indicate rates similar to those for more conventional potentials of -850mV(Ag/AgCl). Overall, there appears to be little offshore experience of the long term effects of CP potentials close to the free corrosion potential.

Resistor controlled CP has been proposed for internal CP of stainless steel seawater piping with the main aim of reducing anode consumption [8.13]. The method employs normal anodes connected via a resistor that stops the full anode potential being realised in applications with low current requirements. The method does not have the required level of control that would be required for a variable higher current requirement material such as high strength steel in marine environments.

In response to the need for low voltage sacrificial anodes, new aluminium alloy anodes [8.14] (Al-0.1%Ga) are being developed. Initial studies have shown that they achieve potentials in the range -773 to -803mV(Ag/AgCl)[3] in sterile seawater. Other work [8.15] attached the anodes to fracture mechanics specimens buried in a coastal sediment. After 190 days the specimens were raised and the potential measured (while still suspended in open seawater) was found to be in the range -680 to -815mV(Ag/AgCl). Whilst more development is required, if proved in service such anodes could have a significant benefit offshore.

Alternatively, modification to the design of the jack-up can be made so that the high strength steel is not exposed to environments where significant hydrogen absorption would be expected, e.g. anaerobic sediments. For example, by utilising a concrete base tank to the structure or by using a lower strength steel with less susceptibility to hydrogen cracking in piled sections of the structure. The cracking discovered in the late 1980s was frequently associated with anaerobic conditions within the spud can and local anodes delivering high negative voltages. The use of an elevated base for a production platform eliminates this particular problem since the high strength steel in no longer exposed to potential anaerobic conditions. This potential solution is of cause limited by the suitability of local conditions.

8.4 SUMMARY

Field surveys found cracking in drilling jack-ups, in some cases irrespective of rig type. Remedial measures of removing anodes from spud cans, the use of biocides and voltage limiting diodes have reduced the incidence of cracking. However, in most cases where cracking occurred the level of CP was not measured unfortunately. Hence, there remained some doubt as to its causes and hydrogen induced cracking remained a serious possibility in several cases. It appears that the increased hydrogen embrittlement susceptibilities of some high strength steels will mean that operators will need to maintain a more positive CP potential than previously used, even if this means accepting a limited amount of corrosion of the structure. The choice of CP system is heavily dependent on location, material and loading. A well modelled CP system incorporating voltage limiting diodes and coatings can work very well but the reported problems are of concern. At present, the successful

---

[3] Laboratory studies often use a saturated calomel electrode (SCE). For consistency these values have been expressed with respect to a Ag/AgCl reference electrode, i.e. minus 22mV (SCE is 244mV).

development of low voltage sacrificial anodes would appear to offer the most satisfactory long term solution. The level of CP for steels with yield strengths >600MPa should be evaluated on an individual basis from HE testing in representative service conditions. In general, CP potentials should be kept more positive than -850mV(Ag/AgCl) unless hydrogen embrittlement test data clearly demonstrate that more negative potentials are not damaging, and should be monitored closely.

REFERENCES

8.01 Marine Technology Directorate Ltd – Design and operational guidance on cathodic protection of offshore structures, subsea installations and pipelines' MTD Report 90/102 (1990)

8.02 *Wanklyn J N* – 'Input data for modelling marine cathodic protection' In Cathodic Protection – Theory and Practice. Eds. Ashworth V and Booker CJL, Ellis Horwood, pp68-77 (1986)

8.03 *Batt C L and Robinson M J* – 'Optimizing cathodic protection requirements for high strength steels under marine biofilms' Corrosion Management, Number 31, September/October (1999)

8.04 *Batt C and Robinson M J* – 'Cathodic Protection Requirements for High Strength Steel in Seawater assessed by Potentiostatic Weight Loss Measurements', British Corrosion Journal, 37 (1) (2002)

8.05 *Thomason W H* – 'Quantitative Measurement of Hydrogen Charging Rates Into Steel Surfaces Exposed to Seawater Under Varying Cathodic Protection Levels' Materials Performance July p267 (1988)

8.06 *Robinson M J and Kilgallon P J* – A review of the effects of sulphate reducing bacteria in the marine environment on the corrosion fatigue and hydrogen embrittlement of high strength steels' Health and Safety Executive, Offshore Technology Report OTH 98 555, HMSO

8.07 *Barker J C* – 'Data surveys of hydrogen assisted cracking in high strength jack-up steels' Offshore Technology Report OTO97 072, Health and Safety Executive (1997)

8.08 *Abertheny K, Fowler C M, Jacob R and Davey V S* – 'Hydrogen cracking of legs and spudcans on Jack-up drilling rigs – A summary of results of an investigation' Offshore Technology Report OTH91 351, HSE Books, Sudbury, Suffolk (1993)

8.09 *Kilgallon P J* – 'The effect of sulphate reducing bacteria on the hydrogen absorption of cathodically protected high strength low alloy steels' PhD thesis, Cranfield University, 1994

8.10 *Davey V S* – 'The cathodic protection of high strength steel in a marine environment' Offshore Technology Report OTN 1999 015

8.11 *Wolfson S L* – 'Corrosion control of subsea piping systems using thermal sprayed aluminium coatings' Materials Performance 35 (7) pp32-36 (1996)

8.12 *Robin Jacob* – Personal communication

8.13 *Johnsen R, Gartland P O, Valen S and Drugli J M* – Internal cathodic protection of seawater piping by the RCP method. Materials Performance 35 (7) pp17-21(1996)

8.14 *Pautasso J P, Le Guyader H and Debout V* – 'Low voltage cathodic protection for high strength steels: Part 1 – Definition of a new aluminium galvanic anode material' Paper 725, Proceedings of Conference Corrosion '98, 22-27 March 1998, San Diego, California, NACE, pp725/1-725/27

8.15 *Batt C L* – 'Optimising cathodic protection requirements for high strength steel in the marine environment' PhD thesis, Cranfield University, 2000

recombination of atomic hydrogen to form hydrogen molecules (and eventually hydrogen gas bubbles) which effectively keeps atomic hydrogen on the metal surface for a longer period resulting in a larger proportion being absorbed by the metal. The hydrogen uptake can increase by as much as an order of magnitude compared to that in sterile conditions at the same CP potential [9.03, 9.04].

When sulphides are present, as in sour oil and gas environments, it is often termed sulphide stress corrosion cracking (SSCC). In both cases, these are forms of HE. They are quite distinct from hydrogen induced cracking (HIC), usually associated with pipeline steels, in which hydrogen generated by internal corrosion in sour oil or gas is absorbed by the steel and collects at elongated manganese sulphide inclusions where it leads to stepwise cracking.

9.3 RECENT LITERATURE REVIEW

The review [9.05] intended to clarify some of the conflicting information about HE resistance that exists at present and provide an explanation of why some steels in the same strength range seem to be inherently more resistant to embrittlement than others. The review considered the HE of offshore constructional steels with yield strengths of ≥ 450MPa. The review showed that the embrittlement behaviour of offshore steels is broadly comparable to that of other steels within the same strength range. The following conclusions and recommendations were made;

1. Microstructure has a controlling influence on HE susceptibility and is a more significant factor than the alloy composition. The susceptibilities of different microstructures can be ranked in the order: tempered martensite < tempered bainite < spheroidised ferrite and pearlite < coarse ferrite and pearlite. Untempered martensite, that may be present in the heat affected zone (HAZ) of welds, is generally regarded as having the highest susceptibility.
2. The principles for the microstructural control of HE have been defined. The way in which hydrogen is trapped by features in the microstructure is particularly important. The traps should be sufficiently numerous and they should also be irreversible so that the hydrogen is held innocuously. The optimum resistance can be achieved with a structure containing fine carbide particles, uniformly distributed throughout the material, which trap large quantities of hydrogen and prevent local concentrations from reaching the level required to cause decohesion and crack propagation.
3. Tempered martensite has been shown to be the preferred microstructure. The steel should have a fine prior austenite grain size and sufficient hardenability to produce 100% martensite through the material thickness. Microalloying and appropriate heat treatments can be used to produce the required size and distribution of precipitates to act as effective hydrogen traps. Tempering at high temperatures is beneficial in reducing the density of dislocations that are reversible hydrogen traps and increase the solubility of hydrogen in the steel. Good tempering resistance is required to avoid the growth of the carbide particles at high temperatures.
4. In the case of offshore steels, recent developments in alloy chemistry and fabrication processes have resulted in alloys with improved mechanical properties and lower carbon equivalent values. This has had the beneficial effect of increasing the strength, particularly for quenched and tempered steels, without necessarily increasing HE susceptibility.
5. The strength or hardness of a particular grade of steel can give a useful first indication of its likely HE susceptibility. However, it has been shown that when steels of different grades are considered together, the strength or hardness is poorly correlated with HE susceptibility. It is concluded that HE susceptibility is more sensitive to the specific nature of the microstructure than to the strength level of the material.
6. By controlling the composition and microstructure some modern HSLA offshore steels have been produced which have HE susceptibilities that are as low as those of BS4360 Grade 50D steels and significantly lower than would be expected on the basis of their higher strength alone.
7. However, HSLA offshore steels exhibit considerable variability in their susceptibilities to HE. It is recommended that each steel should be considered individually and should be subjected to thorough testing before being accepted for use, particularly in critical locations and in

circumstances that could lead to hydrogen charging. This is especially important in the case of steels for the construction of fixed jack-up platforms which are intended to remain in position for the life of the field and where inspection is more difficult to carry out.

8. The effects of hydrogen charging from CP should be fully considered, particularly if there is a possibility that over protection may occur. The use of potential limiting diodes or sacrificial anodes with a less protective potential would reduce the risk of hydrogen assisted cracking occurring in susceptible steels.
9. Sulphides generated by microbial activity in the marine environment can cause a substantial increase in hydrogen uptake by freely corroding and cathodically protected steel. It is recommended that when high strength steels are to be used in conditions where they may be exposed to microbial activity they should be evaluated first using test environments containing similar sulphide levels.

9.4 EFFECT OF WELDING

The increased HE susceptibility in the HAZ of welds is a result of the microstructural changes that occur during the thermal cycle, particularly the formation of martensite, which may not be adequately tempered by subsequent welding passes. Twinned martensite has a particularly high susceptibility to HE. The related welding difficulty of cold cracking, which is a form of HE caused by hydrogen absorbed during welding, results from the extreme susceptibility of twinned martensite in the HAZ of medium carbon steels. Low carbon weldable steels have been developed to avoid this problem.

9.5 HE TESTING

Three basic approaches can be used to determine the HE behaviour of steels [9.06]:

1. The use of smooth specimens and static loads to generate time-to-failure and threshold stress data.
2. The use of smooth specimens, combined with monotonic tensile loading, as in slow strain rate testing (SSRT).
3. The use of pre-cracked specimens in conjunction with fracture mechanics concepts to determine threshold stress intensity values and to generate curves of the time-based crack growth rate, da/dt, against stress intensity, K.

These test types are not equivalent and each can have a specific role. For example, some tests will indicate the fitness for service of a specific material in the given environment, whilst others are used to rank the performance and obviously fracture mechanics based tests are applicable for a fracture mechanics design approach. Commonly used tests are uniaxial tensile testing (smooth tensile specimen), four point bend testing, C-ring testing, double cantilever beam testing (DCB) and SSRT.

A discussion of some of the practical problems involved with HE testing is given in appendix 9.

9.6 HYDROGEN EMBRITTLEMENT TEST RESULTS

Table 9.1 gives SSRT results for a BS4360:50D type parent plate material tested in air and at different CP potentials [9.07]. The effect of strain rate is demonstrated by the %RA being lower at the slower strain rate as more time was available during the test for hydrogen uptake and embrittlement to occur. Table 9.1 also gives results for BS4360:50D steels that had been heat treated to simulate the microstructure in the HAZ of a weld [9.07]. It is clear that BS4360:50D steels do exhibit some embrittlement, even at low levels of CP. This is due to the severity of SSRT and does not necessarily reflect the behaviour of the material in service. It is also apparent that BS4360:50D steels exhibit a wide range of embrittlement behaviour, depending on their composition and the experimental conditions employed, particularly the strain rate.

The results of SSRT of two Q&T high strength offshore steels, SE500 (500MPa yield strength) and SE700 (700MPa yield strength), are shown in Figure 9.1 and are compared with results for E36 steel (described by the authors as a typical BS4360:50D equivalent steel, but having a somewhat unusual composition) [9.08]. Again, the effect of strain rate is seen with a higher embrittlement index[4] (EI) measured in tests conducted at a lower strain rate. When tested in seawater the Q&T steels performed slightly better than the BS4360:50D equivalent steel. Even in a saturated H_2S solution, the 700MPa steel had a similar EI to the BS4360:50D equivalent steel. Clearly, this level of embrittlement would not be acceptable in practice (EI values above 0.6 are sometimes considered to indicate substantial embrittlement) but it is argued that saturated H_2S provides an accelerated test condition for comparing the properties of different materials. It would be concluded from these results that the SE700 would be resistant to HE in seawater by virtue of the similarity in behaviour to the BS4360:50D equivalent steel.

Figure 9.2 gives more SSRT results for the same steels, showing the performance in seawater containing very low levels of H_2S [9.08]. The EIs were lower than those in saturated H_2S but in this case the 700MPa steel was appreciably more susceptible than either of the two lower strength steels. As low levels of H_2S could arise from microbial activity in the marine environment (100ppm has been suggested [9.03]) this could be an important difference and the 700MPa steel could be at risk of cracking in marine applications. This example illustrates the essential requirement for embrittlement testing of high strength steels to be carried out in environments that closely resemble the expected service conditions.

Results reported by Pircher [9.09] also demonstrate that Q&T steels can be more resistant to embrittlement than BS4360:50D as shown in Figure 9.3 which shows SSRT results for two normalised and two Q&T steels tested in natural seawater with CP. The 355MPa yield strength normalised steel was a BS4360:50D equivalent. The Q&T 690MPa steel had similar performance to the BS4360:50D equivalent steel and the Q&T 500MPa steel was more resistant. The results indicate the superior embrittlement resistance that can be achieved with a Q&T microstructure.

Cole [9.10, 9.11] compared the HE of two high strength steels (1%CrMoNb and $3^1/_2$%NiCrMoV) and one medium strength steel ($2^1/_4$%CrMo) that could be considered for marine applications. Three point bend tests were carried out in natural seawater with the addition of approximately 250ppm H_2S to represent the effects of microbial activity. The specimens were in the as-received condition or heat treated to simulate the microstructures that would result from welding. The low threshold stress intensity factor, K_{th}, values found for the HAZ microstructures indicated that they would be particularly susceptible to embrittlement due to the presence of untempered martensite. Cole concluded that in the Q&T condition the two high strength steels performed as well as the medium strength steel and increasing the yield strength does not necessarily increase susceptibility to embrittlement.

Robinson & Kilgallon [9.04] measured K_{th} values for two Q&T offshore steels, SE500 (500MPa yield strength) and DES690 (690MPa yield strength) that had been heat treated to give martensitic microstructures, simulating the HAZ of welds. After heat treatment the hardness values were 323 and 425Hv, which corresponded to yield strengths of approximately 750 and 1030MPa respectively. Although these steels contained the same microstructural phases, there were marked differences in their embrittlement susceptibilities. At a potential of -830mV(Ag/AgCl) in seawater containing approximately 200ppm of SRB generated sulphide, the K_{th} values of the 500 and 690MPa steels were 16 and 39MNm$^{-3/2}$ respectively. Lowering the potential to -1030mV(Ag/AgCl) reduced these values to 10 and 35MNm$^{-3/2}$ as shown in Figure 9.4.

[4] Embrittlement index is (1-%$RA_{environment}$)/%RA_{air}

Batt and Robinson [9.12] carried out a similar study but tested a 900MPa Q&T steel intended for marine applications with a tempered martensitic microstructure and a hardness of 375 Hv. Although the hardness of this steel was lower than that of the DSE690 simulated HAZ described above, it was more susceptible to HE as shown in Figure 9.4. HE occurred in the parent plate in natural seawater, even at -830mV(Ag/AgCl), and the K_{th} values were lower than those for the DSE690 simulated HAZ in seawater at all potentials. The high susceptibility of this steel is particularly significant when it is considered that its microstructure consisted of tempered martensite, in contrast to the untempered martensite of the weld simulations in the other two steels. The high nickel content of this steel (5%) may have been a factor in its poor performance, although there is conflicting evidence for the embrittling effect of nickel additions.

9.7 SUMMARY

It is clear that high strength steels are susceptible to HE when cathodically overprotected particularly in sulphide containing environments. This has been demonstrated using tests such as SSRT where a loss of ductility is found compared to air tests. But SSRT is a severe test (to failure) and does not accurately represent the conditions found offshore. In addition, conventional 350MPa steels also show HE in SSRT and these materials have no history of HE problems. Yield strength and hardness give some indication of the HE susceptibility of a particular steel but as microstructure is also important these measures can only give an approximate guide. More data are required for both the conventional 350MPa steels and the higher strength steels. The HE susceptibility of individual steels will have to be assessed using the most appropriate test type. It would be helpful for all the test types but particularly SSRT to be standardised to conditions most applicable to offshore use to allow comparison and assessment of scatter. It would also be valuable to know if there are any relationships between SSRT and fracture mechanics test results. There are however many parameters which are far from being widely agreed upon, such as precharging effects, sulphide levels and strain rate.

REFERENCES

9.01 *Abertheny K, Fowler C M, Jacob R and Davey V S* – 'Hydrogen cracking of legs and spudcans on Jack-up drilling rigs – A summary of results of an investigation' Offshore Technology Report OTH91 351, HSE Books, Sudbury, Suffolk (1993)

9.02 *Barker J C* – 'Data surveys of hydrogen assisted cracking in high strength jack-up steels' Offshore Technology Report OTO97 072, Health and Safety Executive (1997)

9.03 *Robinson M J and Kilgallon P J* – A review of the effects of sulphate reducing bacteria in the marine environment on the corrosion fatigue and HE of high strength steels' Health and Safety Executive, Offshore Technology Report OTH 98 555, HMSO

9.04 *Robinson M J and Kilgallon P J* – 'HE of cathodically protected HSLA steels in the presence of sulphate reducing bacteria' Corrosion 50 (8) pp620-635 (1994)

9.05 *Robinson M J and Kilgallon P J* – 'A review of the effects of Microstructure on the HE of high strength offshore steels' Health and safety executive, Offshore Technology Report OTO 99 056, HMSO

9.06 EFC – 'Guidelines on materials requirements for carbon and low alloy steels for H_2S-containing environments in oil and gas production' European Federation of Corrosion Publications Number 16, Institute of Materials (1995)

9.07 Unpublished results

9.08 *Coudreuse L, Renaudin C, Bocquet P and Cadiou L* - Evaluation of Hydrogen Assisted Cracking Resistance of High Strength Jack-up Steels. Marine Structures 10 (2-4) pp85-106 (1997)

9.09 *Pircher H, Drodten P, Kaiser H-J and Musgen B* – Behaviour of high-strength structural steels exposed to seawater. Thyssen Technische Berichte, Heft 1/90, pp53-67 (1990)

9.10 *Cole I S and Brook R* – Fractography of crack growth in stress corrosion cracking and corrosion fatigue growth in a number of offshore steels. Proceedings 3rd International Offshore and Polar Engineering Conference, Singapore, pp296-302 (1993)

9.11 *Cole I S, Brook R and Howard I* – A comparative study of the performance of three steels for offshore applications. Proc. Conf. Fatigue of Offshore Structures, London, September, pp167-184 (1988)

9.12 *Batt C L and Robinson M J* – Optimizing cathodic protection requirements for high strength steels under marine biofilms. Paper 13 Proc. UK Corrosion '98, 20-21 October, pp1-5 (1998)

9.13 *Boellinghaus Th, Hoffmeister H and Dangeleit A* - Scatterband for hydrogen diffusion coefficients in micro-alloyed and low carbon structural steels. Welding in the World 35 (2) pp83-96 (1995)

9.14 *Gangloff P R* - A Review and Analysis of the Threshold for Hydrogen Environment Embrittlement of Steel. In Corrosion Prevention and Control. Proceedings 33rd Sagamore Army Materials Research Conference. Ed. Levy, M. & Isserow, S. US Army Laboratory Command. Watertown, MA. 1986

Table 9.1 SSRT results for parent BS4360:50D type steel and simulated HAZ (from 9.07)

Condition	Environment (Ag/AgCl)	Strain Rate (s^{-1})	RA (%)
Parent	Air	9.3×10^{-6}	83
Parent	Seawater + CP = -800	3.7×10^{-7}	78
Parent	Seawater + CP = -850	3.7×10^{-7}	57
Parent	Air	2.5×10^{-5}	81
Parent	Seawater + CP = -1020	2.0×10^{-6}	53
Parent	Seawater + CP = -1020	5.0×10^{-7}	39
Sim. HAZ	Air	9.3×10^{-6}	71
Sim. HAZ	Seawater + CP = -850	3.7×10^{-7}	29
Sim. HAZ	Air	2.5×10^{-5}	67
Sim. HAZ	Seawater + CP = -850	5.0×10^{-7}	45
Sim. HAZ	Seawater + CP = -1000	5.0×10^{-7}	13
Sim. HAZ	Air	1.0×10^{-6}	50
Sim. HAZ	Seawater + CP = -850	1.0×10^{-6}	27
Sim. HAZ	Seawater + CP = -950	1.0×10^{-6}	19
Sim. HAZ	Seawater + CP = -1050	1.0×10^{-6}	19

Figure 9.1 – EI for SSRT on E36 (BS4360:50D equivalent) and two high strength Q & T steels showing effects of strain rate and sulphide [9.08]

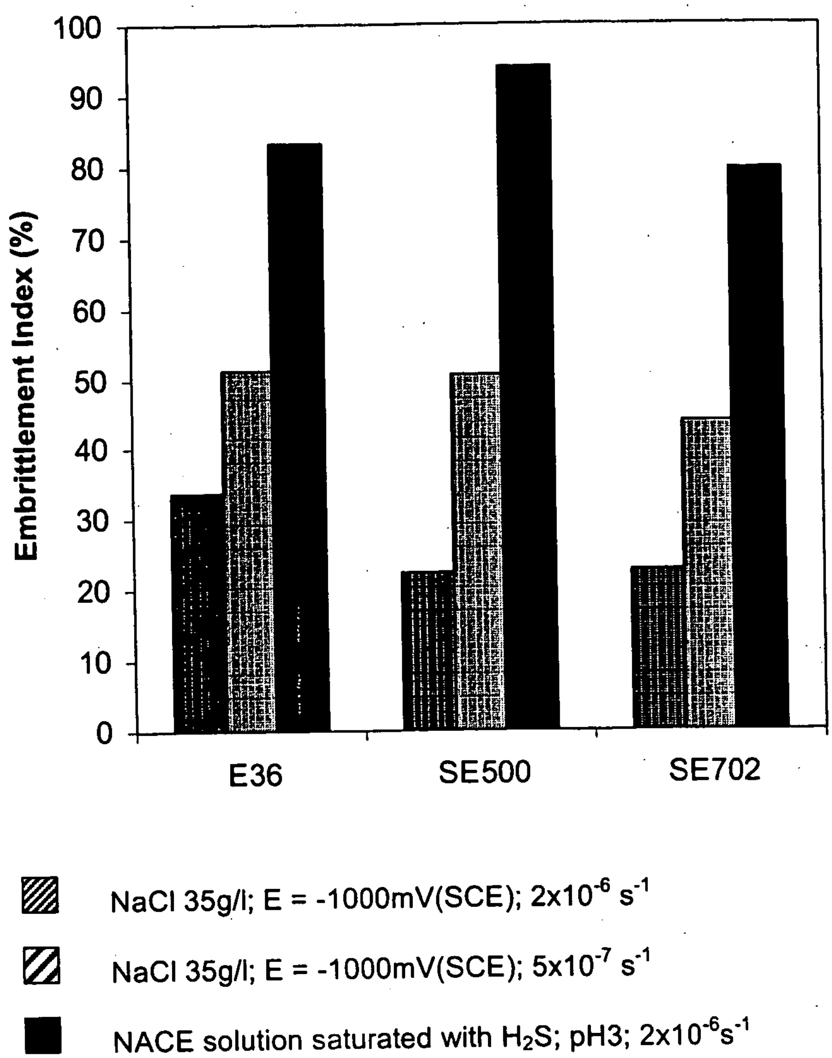


Figure 9.2 – EI for SSRT tests showing the effect of trace quantities of H_2S [9.08]

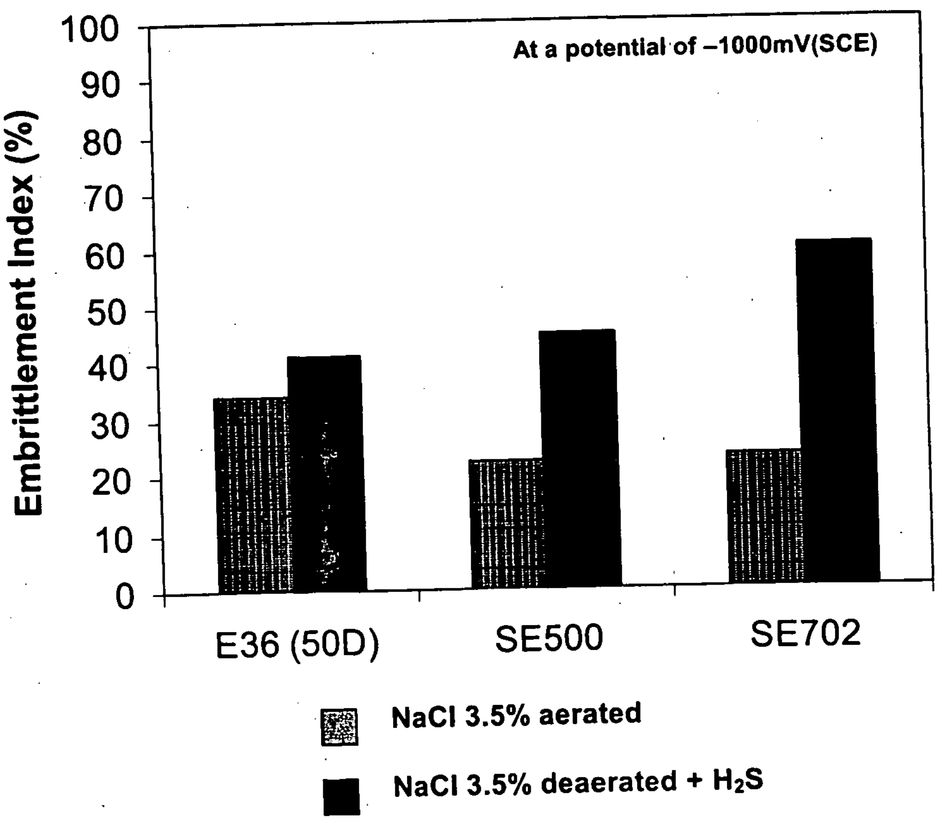


Figure 9.3 – Percentage reduction in area in SSRT showing the effect of applied potential in seawater [9.09]

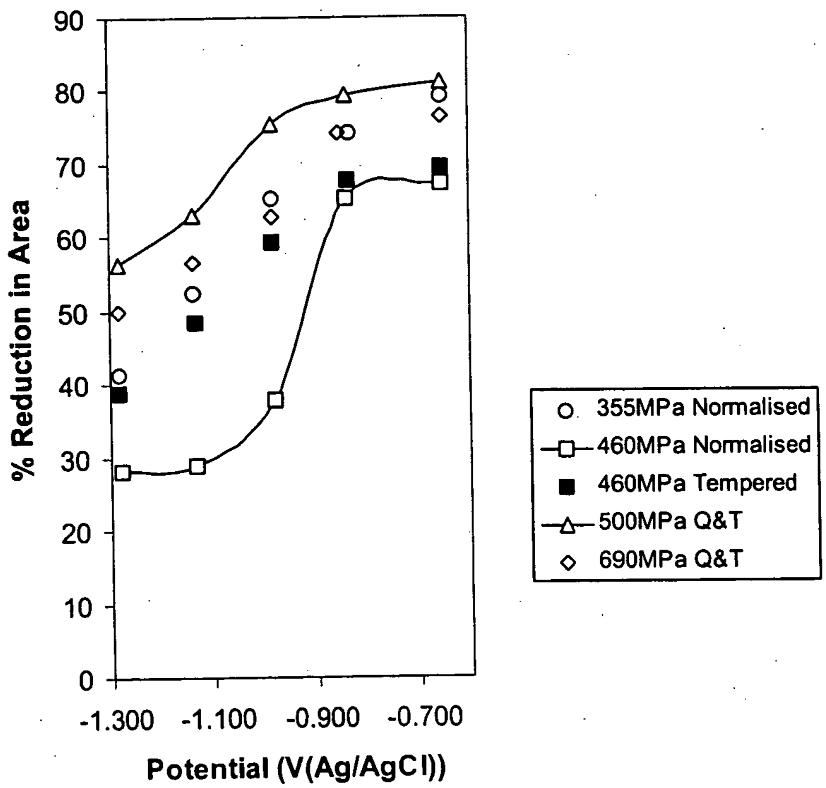
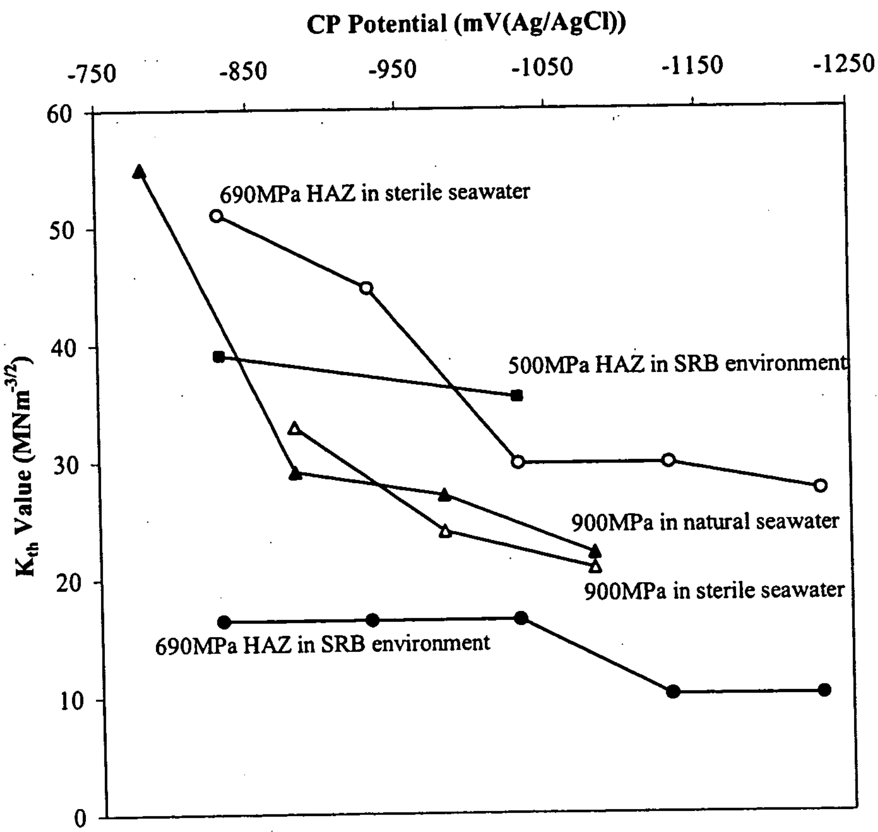


Figure 9.4 – Comparison of K_{th} values for 900MPa steel, 500MPa steel simulated HAZ and 690MPa steel simulated HAZ in sterile seawater, natural seawater and seawater containing SRB with applied cathodic protection [9.03, 9.12]



10. HIGH TEMPERATURE PROPERTIES

Steel offshore installations can be subjected to high temperatures as a result of fire, either a pool or jet fire, or possibly fire on the sea. The resulting loss in strength can lead to partial or total collapse of a leg. Local deformation can also impinge on critical equipment. An important safety requirement is to maintain a sufficient time for evacuation of personnel. Hydrocarbon fires are recognised as more damaging than cellulosic, resulting in higher temperatures (up to ~1100°C after a few minutes). Fire is considered an accidental load in most standards and codes. In this case permanent deformations are allowed, provided that they are not excessive to threaten the integrity of the installation. Acceptance criteria are normally based on retention of strength during the fire for an adequate period of time, which is normally defined in terms of a limiting temperature, or a limiting strain or deflection.

In terms of temperature the acceptance criterion requires the temperature of the structural steel to be limited to a given value, typically in the range 400 - 500°C. This is normally the temperature at which steel exhibits approximately a 50% reduction in yield stress [10.01]. This simple approach is based on several assumptions including that the structure heats up uniformly and that differential heating does not influence the behaviour of the structure. In practice differential heating can lead to local high stresses, depending on the restraint exerted on heated members by the surrounding cooler structure. In addition, the buckling capacity of tubulars is dependent on temperature, through the effect on Young's Modulus.

BS 5950 Part.8 also provides stress reduction factors at elevated temperatures, as well as critical temperatures (related to performance fall-off), for medium grade structural steels. The reduction in strength with temperature is given at selected strains. The appropriate strain for columns in compression is 0.5%, whereas for members in tension it is 1.5% strain.

Two test methods are in use to produce suitable design data (i) steady state heating, (ii) transient test method (heating at constant rate, typically 10°C /min) under stress. BS 5950 covers (i), but for (ii) which is more typical of practice, there is no test standard published.

Protection of steel structural members is normally by using passive fire protection in the form of coatings. These delay the rise in temperature to critical levels, reducing the risk of escalation and providing time for evacuation of personnel. PFP coatings are normally either cementitious, intuminescent or refractory fibres. Such materials can be applied to high strength steels in the same way that they are applied to medium strength steels.

Most available test data, however, refer to loss of strength in 50D type steels tested as isolated components and there appears to be little published data for the performance of modern high strength steels (with yield strengths > 500MPa) under fire conditions. As part of the first phase of the joint industry project on Fire & Blast, coordinated by the Steel Construction Institute (SCI), a review was undertaken of the experimental data relating to the performance of steel components at elevated temperatures [10.03]. The steels with the highest yield strength included in this were RQT 501 (approximately 550MPa). This is a significant omission in performance data for higher strength steels.

More recently data have been published for the high temperature performance of both Q&T and thermo-mechanically rolled steels with yield strengths ~ 450MPa, for a range of thicknesses [10.04, 10.05]. This includes reviewing previously published data as well as reporting on new tests on medium and higher strength steels. The latter includes 450MPa steels in the quench and tempered condition, of three thicknesses (10, 40 and 60mm plate). Strength factors were derived for various limiting strains, ranging from 0.5 – 5%. For all thicknesses 50% loss of strength occurred between 550°C and 600°C. Comparison of strength factors with those given in BS 5950 part 8 showed that actual performance for the strength loss at the BS specified limit of 0.5% strain was equal to, or better than, the design curves (see Figure 10.1). However, there is evidence that at temperatures in excess of

650°C Grade 450 steels do appear to deteriorate at a slightly faster rate than lower strength steels, considered to be due to additional softening from over-tempering.

A parallel study of Grade 355 TMCR and normalised steels showed that these steels exhibit properties which are well below those given in the BS standard, particularly at 0.5% strain. For normalised 355 steels the elevated temperature strength was lower than that of the 450 steel. It was also shown that compositional changes in different thicknesses of grade 450 EM(Z) did not lead to significantly different properties at elevated temperatures in terms of fire performance.

Overall, the strength of all steels will decrease with increasing temperature and at very high temperatures most steels will show a similar performance. Since high strength steels can have different compositions and can obtain their improved strength from a number of different manufacturing routes it is likely that their performance at intermediate temperatures will differ significantly. Hence data are needed for each material type, as demonstrated by the recently published data for 355 & 455 grade steels. This is demonstrated by data from both 355MPa and 450MPa steels showing significantly different high temperature performance, with the 450MPa steel having better relative properties at temperature. These differences may be due to, for example, grain refinement, which is employed to increase strength. Grain coarsening through exposure to high temperatures will lead to reduced strength. Quenching and tempering is also one of the standard processes for achieving high strength. The temperature range for tempering is usually 580 - 620°C and further excursions to this temperature level and above are likely to reduce strength.

No test is known on structural frameworks with a geometry similar to either tubular joints or jack-up leg, and hence the possible loss in strength of the welded tubular high strength steel joints from increased temperatures is not known.

10.1 SUMMARY

Acceptance criteria for steel structural components under high temperature conditions are normally based on retention of sufficient strength during the fire for an adequate period of time, to allow evacuation, which is normally defined in terms of a limiting temperature, or a limiting strain or deflection. BS 5950.Part 8 [10.2] provides data on these parameters but is limited to lower strength steels. Some data have recently been published for a limited set of steels which shows that the properties of this type of Q&T steel are similar to the steels covered by BS 5950 Part 8, at least up to 650°C. However, for even higher strength steels there appears to be no published data, which is a significant limitation at present, and test data is required to demonstrate satisfactory high temperature performance.

REFERENCES

10.01 FABIG Technical Note, 'Fire Resistant design of Offshore Topside structures', Feb.1993

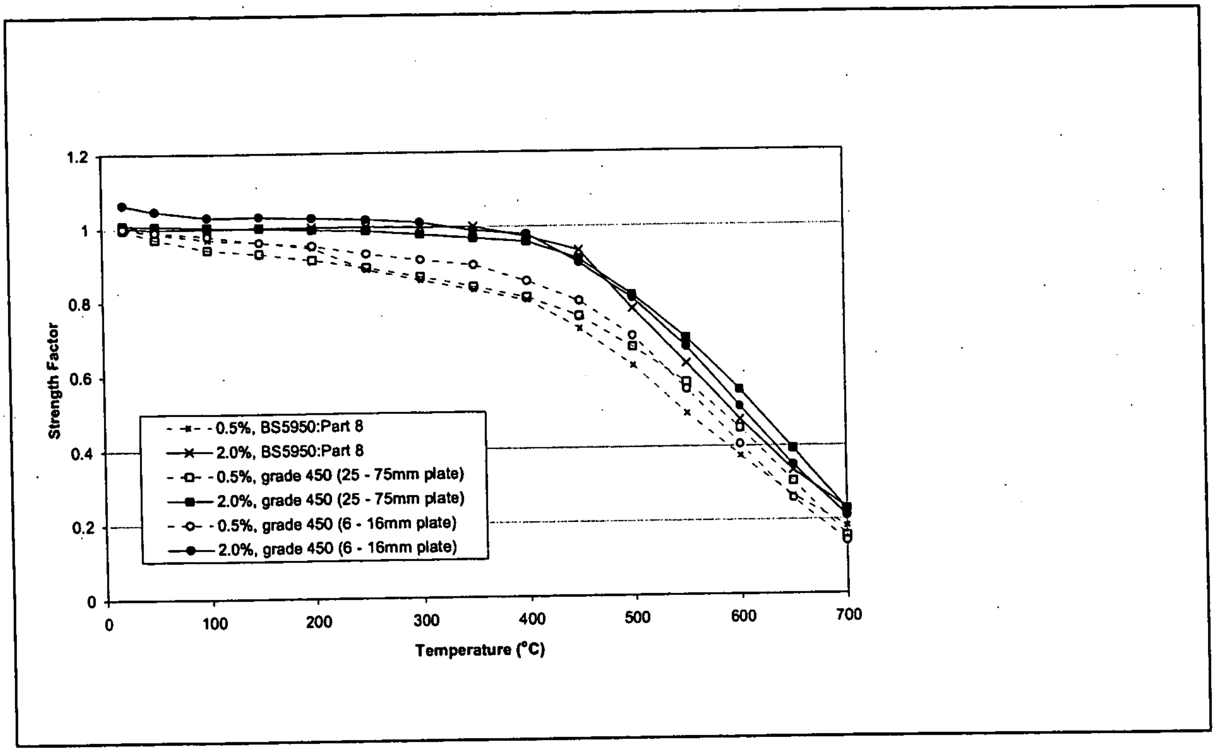
10.02 British Standards Institution, BS 5950, The structural use of steelwork in buildings, part 8, code of practice for fire resistant design', 1990

10.03 Steel Construction Institute, 'Experimental data relating to the performance of steel components at elevated temperatures', Offshore Technology report OTI 92 602, HSE Books, 1992

10.04 FABIG Technical Meeting Note, 'New Data on Materials for use offshore', 2000

10.05 Steel Construction Institute, 'Elevated temperature and high strain rate properties of offshore steels', Offshore Technology report OTO 020, 2001, HSE Books

Figure 10.1 – Comparison of strength factors between Grade 450 EMZ with those given in BS 5950 Part 8 [10.05]



11. HIGH STRAIN RATES

The properties of ferritic steel are strain rate sensitive and the yield stress increases as the rate of loading is increased. The yield ratio (yield stress/ultimate tensile stress) also rises. Fracture toughness therefore is affected by strain rate.

In terms of crack initiation, an increased yield stress promotes brittle fracture by default. The ductile-to-brittle transition is moved to higher temperatures. Until 1987, K_{IC} testing was performed at a nominally static loading rate, usually defined in terms of the rate of increase in the applied stress intensity factor K_{app}. The British Standard BS 6729:1987 'Determination of the dynamic fracture toughness of metallic materials' [11.01] extended the procedures to rates of increase in K_{app} of 10^5 $N.mm^{-3/2}.s^{-1}$ (= 3160 MN. $m^{-3/2}.s^{-1}$) and in CTOD of up to 150mmThe fracture toughness under these dynamic load applications is still termed K_{IC} and for most materials the value is below the static fracture toughness. Such values will relate to impact loading, such as collisions and dropped objects, but do not extend to blast loading from explosions.

With dynamic conditions, crack propagation and crack arrest may also need to be considered. During the propagation phase, a certain stress intensity needs to be applied to keep the crack in motion and this is usually termed K_{ID}. Its value depends on the crack velocity but the dynamic toughness is relatively insensitive to velocity, except when the crack speed approaches the limiting speed in the material [11.02; 11.03].

The effect of the rate of strain on the mechanical properties of offshore steel has been studied by Webster [11.04] and by the Steel Construction Institute [11.05]. In general, offshore strain rates vary from 10^{-4} s^{-1} for wave loading, 10^{-2} to 10^{+1} s^{-1} for ship collisions, and anything from 22 to 10^{+6} s^{-1} for blast effects. The tests on 355MPa and 450MPa steels in [11.04] showed steady increases in lower yield stress (LYS) and the stress for 5% plastic strain as the strain rate increased from 10^{-3} to 10^{+1} s^{-1}, but this increase was only a modest 16% for each parameter. It was pointed out that the inter-cast variation in the steel is likely to give a greater difference than this. A greater response was observed for the upper yield stress (UYS) because of the contribution from solute locking. Using 700MPa pipeline steel, a steeper rise in proof stress and UTS was observed at strain rates from 0.5 to 45 s^{-1} producing increases of about 50%.

Additional work by British Steel (Corus) as part of the Structural Integrity Assessment Procedures for European Industry (SINTAP) [11.06] investigated the strain-rate corrected fracture toughness determined from Charpy impact energy [11.05]. For strain rates from 10^{-4} to 10^{-1} s^{-1} the transition temperature was predicted to rise by about 40°C for 355MPa yield stress steels but this reduced to about 5°C in one model and below 20°C in another when the yield stress was increased to 1000MPa. This supports the view that the strain rate sensitivity of steels is less significant as the strength rises.

The variations in yield strength with strain rate for three weld metals with yield strengths from 415MPa to 825MPa have been correlated with fracture toughness test results, in a paper suggesting the use of a single parameter to express the combined effects of strain-rate and temperature [11.06]. Values were computed for strain rates from approximately 10^{-3} to 1.5 x $10^{+3}s^{-1}$. Although not highlighted in the paper, the increase in YS was similar for all three materials (about +300MPa) and consequently the percentage change for the high strength weld metal was appreciably lower.

Recently reported work by the Steel Construction Institute [11.06] included new tests on three thicknesses of 450MPa Q & T steels tested at strain rates from 0.001/sec to 10/sec. These showed that all tensile properties increased with increasing strain rate but the upper yield strength increased more rapidly. The increases in tensile properties were greatest for the 60mm plate thickness, with the 400mm plates showing the lowest level of increase. The lower yield strength (LYS) for the 60mm plate showed an increase of ~30% over the range of strain rates tested, whereas the increase in the YS

was higher at ~37%. The measured change in UTS was ~22%. A general equation was developed of the form:

$$\sigma = \kappa\epsilon^{n}(d\epsilon/dt)^{m} + j$$

where
j is the elastic limit stress (i.e. a value of 400MPa was found to give the best fit)
κ is the true stress (MPa at a true strain of 1)
n is the strain hardening exponent
ϵ is the proof strain
$d\epsilon$is the plastic strain rate/sec
m is the strain rate exponent.

Values for these coefficients are given in [11.05] for the 450MPa steel tested. In structural engineering practice total strain is required rather than proof strain and engineering rather than true stress. Conversion factors are given in [11.04].

The changes in stress with increasing strain rate are shown in Figure 11.1.

Although data currently appear to be relatively scarce in this area, it appears from the limited information available that changes in the strain rate applied to steels and weld metals can result in appreciable variation in mechanical properties and this needs to be taken into consideration in safety assessments where high strain rates are applicable. Unfortunately there are few data for steels with yield strengths >500MPa and hence test data are required for each specific application.

11.1 SUMMARY

Offshore strain rates vary from $10^{-4}s^{-1}$ for wave loading, 10^{-2} to $10^{+1}s^{-1}$ for ship collisions, and up to $10^{+6}s^{-1}$ for blast effects. The properties of ferritic steel are strain rate sensitive and the yield stress and the yield ratio (yield stress/ultimate tensile stress) increase as the rate of loading is increased. Strain rate also has an effect therefore on fracture toughness. Data for the effect of strain rate on high strength steels are limited, particularly for steel with YS >500MPa. Recent tests on 450MPa steels showed that all tensile properties increased with increasing strain rate but the lower yield strength (LYS) (for the 60mm thickness plate) over the range of strain rates tested (0.001/sec to 10/sec), whereas the increase in the UTS was somewhat higher at ~37%. The measured change in UTS was ~22%.

Overall, it appears from the limited information available that changes in the strain rate applied to steels and weld metals can result in appreciable variation in mechanical properties and this needs to be taken into consideration in safety assessments, where high strain rates are applicable.

REFERENCES

11.01 BS 6729:1987 'Determination of the dynamic fracture toughness of metallic materials', British Standards Institution, London

11.02 *Rosakis A J, and Freund L B,* 'Optical measurement of the plane strain concentration at a crack tip in a ductile steel plate', Journal of Engineering Materials Technology, Vol 104, 1982, pp 115-120

11.03 *Anderson T L,* 'Fracture Mechanics : Fundamentals and Applications', Section 4, 'Dynamic and Time-dependent Fracture', CRC Press, Boca Raton, Florida, 1991

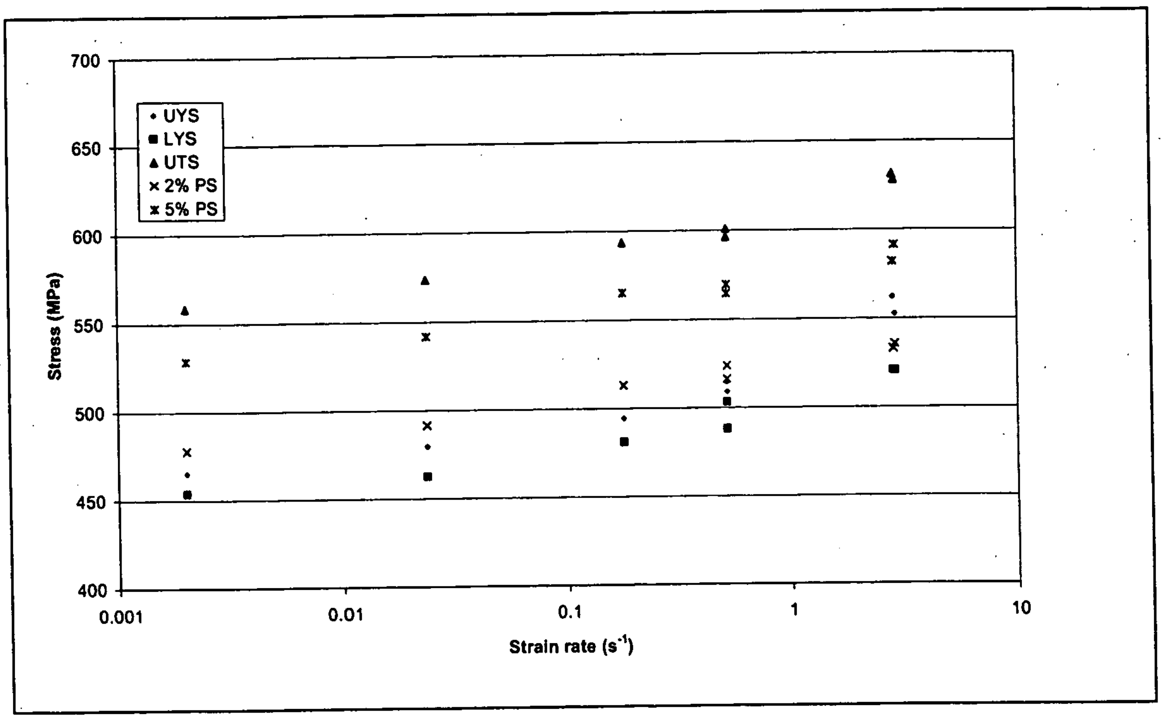
11.04 *Webster S*, 'The Effect of Strain Rate on the Mechanical Properties of Offshore Steels'. In: FABIG Technical Meeting on 'New Data on Materials for Use Offshore', 15-16 May 2000

11.05 Steel Construction Institute, 'Elevated temperature and high strain rate properties of offshore steels', HSE report OTO 020, 2001, HSE Books.

11.06 *Bannister A C*, 'Determination of Fracture Toughness from Charpy Impact Energy: Procedure and Validation', SINTAP Sub-Task 3.3 Report, British Steel plc, January 1998

11.07 *Deans W F and Said M N M*, 'Effect of Loading Rate on Fracture Toughness of Steel Weld Metals', Res Mechanica Vol 28 (1989) pp 251-267]

Figure 11.1 – Influence of strain rate for 60mm Grade 450 EM9Z plate



12. FIELD PERFORMANCE OFFSHORE

12.1 INTRODUCTION

Jack-ups, constructed from high strength steels, have been used for drilling for many decades with generally good performance. These units have been dry docked for inspection and repair on a regular basis. Hydrogen cracking was found in the late 1980s in some jack-ups of this type, which led to a major research programme and guidelines on control of this type of cracking in service.

Production jack-ups have been introduced in recent years - these are on station for many years, without the opportunity for dry dock inspection. BP Harding was installed on the UKCS in 1996, Siri in the Danish sector in 1998, Hang Tuah in the West Natuna Sea Gas development offshore Indonesia in 2001 and Elgin-Franklin in the UK sector in 2001

Since the Hutton tension leg platform was installed in 1984 there have been several other applications of this concept, including Mars and Auger in deeper waters. Some data are available on the in-service performance of the tethers in these TLPs.

This section reviews the design features and field performance of both drilling and production jack-ups, and TLPs to provide feedback on issues that may need attention in future designs.

12.2 PRODUCTION JACKUPS

12.2.1 BP Harding

The BP Harding jack-up is of the TPG 500 design, weighing 23,000 tonnes, was installed in the Harding field in 110m of water, in 1996 [12.01]. It consists of a triangular hull supported by three 125m high legs, on a fixed concrete base. This gravity base weighing 85,000 tonnes also contains storage capacity (570,000 barrels). To improve fatigue life some 600 forged nodes were incorporated in the legs.

The lower sections of each leg are fabricated from ~400MPa steel. The forged nodes were produced from 400MPa Q&T steel. For the chords in seawater and the splash zone, steels with a minimum yield strength of 450MPa were used, whilst in the air chord steels were of higher yield strength (550MPa). The highest strength steel (700MPa) was employed in the racks.

BP Harding has voltage limiting diodes to minimise excessively negative CP levels. It is understood that these have led to more positive potentials in practice (~ 700mv Ag/AgCl) than intended.

12.2.2 Siri

The Siri field is located in 60m of water in the Danish sector of the North Sea. It consists of a three legged jack-up standing on the top of a steel storage tank. The tank was installed in May 1998 and the jack-up installed some six months later. Three papers have been published giving details of Siri [12.02, 12.03, 12.04].

The tubular legs are 104m in length, with an outer diameter of 3.5m and stand 13m deep in the tank structure, the gap between the legs and the sleeves being grouted. The wall thickness of the legs varies from 65 to 110mm. The lower 27m of the legs are without holes for the jacking system and are made of 390MPa steel. The remaining parts of the legs have 460mm diameter jacking holes spaced at 1750mm and are made of high strength steel, with a minimum yield strength of 690MPa. In fact the steel was delivered with an actual yield strength of 800MPa. In some respects this higher strength is beneficial (restricting yielding at the holes where static strength is dominant) but has been found to be a penalty for fatigue and fracture. CTOD tests of the heat affected zone of this high strength steel showed that there were areas of low fracture toughness, and in particular it was found that very little fatigue crack growth could take place before brittle fracture would occur. As a result several studies have been undertaken on the performance of this steel and, although it has been confirmed that the

fatigue life is generally satisfactory, field inspection is necessary at an earlier stage than originally planned. Analysis of the fatigue performance has shown that 8 circumferential leg welds, located at the lower parts of each leg, have fatigue lives less than 200 years.

Inspection of the jack-up legs is complicated by the outer circumferential weld being ground smooth, with no weld cap being visible to identify the location of the weld. The grinding was partly to benefit fatigue performance but also to ensure that the legs could pass easily through the guide rings of the jacking system. The thickness transitions are made on the inner surface of the legs and, as a result of local bending moments, the highest fatigue stresses occur on the inner surfaces, making inspection more difficult. Another complication for inspection is that the outer surface of each leg is coated with hot sprayed aluminium and the inner surface painted.

Cathodic protection of the legs in seawater is provided by both conventional anodes and the sprayed aluminium layer. It is not known at what voltage levels the protection system operates (the potential of sprayed aluminium is in the range -850 to -900mV).

A paper presented at OMAE 2000 [12.04] described an inspection tool for the Siri legs, which combines the benefits of ultrasonic and eddy current methods. This tool is now under development, with both small scale and full scale trials planned to fully evaluate the tool.

12.2.3 Hang Tuah ACE Platform

This provides gas compression facilities for the West Natuna Sea Gas development offshore Indonesia and was installed in 2001 [12.05]. It has a steel gravity base, supporting three legs manufactured in conventional steels and a barge deck. To date, minimal details have been published on materials and structural aspects and in-service experience is very limited to date.

12.2.4 Elgin-Franklin

The production jack-up to be placed in 92m of water in the Elgin field is the TPG 500 design, with an overall weight of 33,000 tonnes. It will be bridge linked to a nearby well-head platform. The triangular hull (9.8m deep) is supported by three legs, each with a spud can fixed by six piles to the seabed. The legs are lattice type, based on three chords spaced at 17.5m centres [12.06].

The steels used in the jack-up legs are Creusot-Loire Superelso E702 and SE500, with the lower strength steel utilised in the lower leg sections close to the seabed. As with the similar jack-up for the Harding field, there is extensive use of forged nodes, each fabricated in half sections which are subsequently welded together.

Cathodic protection is understood to be based on using conventional anodes, with no voltage limiting diodes. The performance of this system has yet to be established in service.

12.3 DRILLING JACK-UPS

As noted earlier, drilling jack-ups have extensive service experience offshore, in a range of water depths.

Drilling jack-ups have suffered many accidents over the years of operation, including foundation problems, ship collisions and fatigue. Fatigue problems have been experienced during dry tow, due to the wind loading on the legs.

HSE has funded a review of field surveys undertaken by Lloyds Register, Det Norske Veritas and ABS of jack-up rigs for damage and cracking [12.07]. The most extensive survey was by ABS which included 89 jack-ups over a 21 year period. The units examined were all 3-leg jack-ups, which had been in service over a period of years commencing between 1975 and 1992. The yield strengths of the

steels were in the range 315 to over 690MPa. The overall review included up to 2000 individual survey reports. Of these, 309 (~15%) had spud can damage or defects noted.

Table 12.1 shows the different steel types used in the different rigs. Most defects were found at the spud cans or on the leg connection to the can. The number of defects found was dependent on the rig design. Overall there were 189 survey reports for the 58 Marathon Le Tourneau rigs surveyed, which contained documented information for over 3000 spud can/leg connection defects. For the 31 Friede & Goldman rigs surveyed, there were 646 spud/can leg connection defects. The only significant difference between the two main types of design was defect number and length, with more defects but a smaller average length (of 150mm) for MLT rigs compared with an average length of 380mm for F&G rigs. This difference was considered to be mainly due to design, with the MLT design consisting of several brackets and shear elements. Trends in damage to the spud can to leg connections were investigated for both in-service time and operating location. The data do not indicate, surprisingly, that operating area has a significant effect on the number and size of defects but the selection of areas was undertaken at a coarse level.

The cause of defects was investigated. A proportion was considered to be due to either fabrication problems, design details or re-cracking after repair. In addition, some defects were considered to be due to extreme loads, cyclic loads or hydrogen cracking, although it proved difficult to differentiate these causes. ABS concluded that the most likely causes of defects not due to design deficiencies were attributed to high global loads arising from moving on location, uneven spud can loading or submerged rocks.

In terms of steel type, most of the cracks reported were found in the Q&T low alloy steels, particularly in the HAZ. In addition, it was found that rigs manufactured at one shipyard in North America had a particularly high number of defects per rig year (the steel used was SS-100).

In terms of hydrogen cracking during or after welding the American Welding Society code determines a 'susceptibly index' (SI), which is based on the PCM [5] and a measure of the hydrogen environment (H = 5,10,15, relating to extra low, low or uncontrolled hydrogen in the welding consumables). From the calculated SI values, certain steels were deemed to be more vulnerable; these included N-30, HY-100, & SSS-100. However, the survey data did not enable these trends to be substantiated.

In terms of in-service performance, hydrogen cracking in twelve jack-up rigs of five different types was discovered in the late 1980s by the Department of Energy [12.08]. Surveys of two CFEM rigs operating on the UKCS identified HAC in the HAZ of weldments at the intersection between leg chords (fabricated with centrifugally cast steel) and the spud can top plate and at some internal connections between bulkhead members and the abutting leg chords. Cracking was then found in other CFEM rigs, including cracking at the external welds between leg chords and the spud can top plates. Severe cracking was found also at the brace to chord connections up to the third horizontal level in another similar rig (at a location where paint had been removed to enable underwater inspection to be carried out). In addition to the CFEM rigs, HAC was found in two Hitachi rigs, in a four legged MSC designed rig and two Marathon Le Tourneau rigs. All of these five rig types showed damage as HAC in the HAZ adjacent to welds.

This problem led to a large research programme on this topic being funded by the Department of Energy. The main conclusions were that the cracking was due to the presence of H_2S in spud cans, too negative cathodic protection levels and poor material selection. As a result, a number of recommendations to control the cracking were made and published in the Department of Energy (more recently HSE) Guidance Notes [12.09]. This included emphasis on materials selection (for example by slow strain rate testing), welding procedures, avoidance of anaerobic conditions and limiting the potentials from cathodic protection to minimise the generation of hydrogen. These

[5] $P_{cm} = C + \frac{Mn+Cr+Cu}{20} + \frac{Si}{30} + \frac{V}{10} + \frac{Mo}{15} + \frac{Ni}{60} + 5B$

REFERENCES

12.01 'A new role for Jack-ups', Offshore Engineer, October 1995

12.02 *Baerheim M, Manschot D, and Nortvedt T,* 'Use of a Jack-up as a permanent installation for the marginal SIRI field', Sixth International Jack-up conference, City University, 1997

12.03 *Baerheim M, Manschot D, Olsen T, and Eide H,* 'The SIRI production jack-up platform', Proceedings of the Offshore Technology Conference, Houston, 1999

12.04 *Chen K, Nichols N, Lalani M, Baerheim B, and Lundberg A,* 'Fatigue & Fracture Considerations in the reliability based inspection planning for the SIRI production jack-up platform', OMAE Conference, New Orleans, 2000.

12.05 *Jackson G, and Pennington D,* 'Installation of the Hang Tuah ACE platform', City University Eighth Jack-up Platform Conference, Sept. 2001, London, published by AST Press, Hove.

12.06 *Thomas, P.A.*' TPG 500 Production Jack-up Design', Sixth International Conference 'Jack-up Platform', City University, Sept. 1997, AST Press, Hove

12.07 *Barke J C,* ' Data Surveys of hydrogen assisted cracking in high strength jack-up steels'. HSE OTO Report 97 072, 1998

12.08 *Abernethy K, Fowler C M, Jacob, R, and Davey V S,* 'Hydrogen cracking of legs and spudcans on jack-up drilling rigs - a summary of results of an investigation', HSE Report OTH 91 351, HSE Books

12.09 *Dept. of Energy/HSE,* 'Offshore Installations, Guidance on Design, Construction, and Certification', 1990, HMSO, London

12.10 *British Steel Company Research Services,* Dept of Energy report OT-R-8241, 1982.

Table 12.1

Steel type	Min.Yield strength/UTS (MPa)	AWS Steel grade	Location	Design
N-30	583/665	G/G+	Rack	Marathon Le Tourneau
N-20BHT (Mo)	480/549	D/E	Chord plates	Marathon Le Tourneau
N-20BHT(Va)	480/549	C-D	Chord plates	Marathon Le Tourneau
C-23M	315/535	D-E	Gussets	Marathon Le Tourneau
M&M tubing	583/686	G	Tubing	Friede & Goldman
HY-100	686/spec	G	Rack chord	Friede & Goldman
SSS-100 Ti	619/686	G	Rack chord	Friede & Goldman
SSS-100- Vn	619/686	G	Rack chord	Friede & Goldman

13. INSPECTION AND REPAIR

Most existing offshore installations have suffered damage and cracking from fatigue or accidental loads, such as ship impact and dropped objects. Such damage is found through regular inspections and repaired as required. However, most inspection and repair techniques have been developed for medium strength steels. Differences in the structural integrity performance of high strength steels would need to be taken into account during inspection and repair. Hence differences in the structural integrity performance of high strength steels would need to be considered during their inspection and repair.

Hydrogen cracking is particularly relevant to high strength steels. Such cracking is generally finer and more heavily branched than fatigue cracking [13.01] and hence may be more difficult to detect and size in the early stages of crack growth, using conventional NDE methods, and removal of any coatings may be necessary. Published work has commented on the lack of data on the reliability of inspection techniques for use underwater, for the detection of hydrogen cracking [13.01]. Visual inspection is only likely to detect large hydrogen cracks, likely to need urgent repair.

The most widely used inspection technique for detecting fatigue cracking in fixed jackets is flooded member detection (FMD) and visual inspection. FMD requires a through thickness crack to develop in a welded joint, enabling an attached member to become flooded. At this stage the remaining life of the joint is limited, both through reduced static strength and limited fatigue life. Several papers have been published assessing the tolerance of structures to through thickness cracking [13.01; 13.02]. These were based mainly on data for medium strength steels. In terms of estimating the residual static strength in joints containing cracks, all of the available data are for medium strength steels. These show that cracks at very early stages of through thickness cracking lead to a ~30% reduction in static capacity, which is generally covered by the normal safety factors. Larger cracks produce a greater loss in capacity, which could be significant in extreme storms, and this demonstrates the need to find through thickness cracking at an early stage. Limited data exists on the effect of cracking in high strength steel joints (SE 702) [13.04], mainly from a series of nine static tests performed on large pre-cracked welded tubular joints. It was shown that the reduction in static strength compared to cracked medium strength steels was about 5% greater. The details of these tests are discussed in section 14. These results suggest that use of FMD for high strength steel joints may require careful assessment of the consequences of extreme wave loading on heavily cracked joints.

Repair of cracked or damaged members of joints has been undertaken using a range of techniques, from underwater welding, grouted or mechanical clamps, or grout filling of members [13.04]. Most cracked joints in jack-ups have been repaired in dry dock using conventional welding procedures. Underwater repair of damaged high strength steel components in production jack-ups may require the development of special techniques.

The use of clamps for repair of high strength steels joints is unlikely to be different from experience with medium strength steels, particularly as the clamp is normally designed to carry the full load across the damaged component. Experience of underwater welding of high strength steel components is very limited, particularly offshore, and hence a large amount of test data and trials are likely to be required to approve the process for use in practice. However, underwater weld repair of X70 pipeline steels has been carried out satisfactorily.

13.1 SUMMARY

Most data relevant to inspection and repair offshore have been developed from medium strength steels, although there is some limited experience in welding high strength steels pipeline materials underwater. Inspection of hydrogen cracking using conventional NDE method is highlighted as being more difficult to detect and size in the early stages of crack growth. There are some limited data for the static capacity of high strength tubular joints containing through thickness cracks relevant to using

flooded member detection. This shows a slightly greater loss of strength than for medium strength steels under extreme wave loading. In practice, underwater repair of steels with yield strengths >500MPa is likely to require some specific test date for that steel before safe offshore application can proceed.

REFERENCES

13.01 *Abernethy K, Fowler C M, Jacob R and Davey V S,* 'Hydrogen cracking of legs and spudcans on jack-up drilling rigs – a summary of results of an investigation', HSE Report OTH 91 351, HSE Books.

13.02 *Sharp J V, Stacey A, and Wignall C M,* 'Structural Integrity Management of Offshore Installations based on Inspection for Through Thickness Cracks', OMAE Conference, Lisbon, 1998.

13.03 *Sharp J V, and Stacey A,* 'Risk Management of Through Thickness Cracks in Offshore Fixed Structures', ISOPE Conference, Brest, June 1999.

13.04 *Abernethy K, Fowler C M, Jacob R, and Davey V S,* 'Hydrogen cracking of legs and spudcans on jack-up drilling rigs – a summary of results of an investigation', HSE Report OTH 91 351, HSE Books.

13.05 *Sharp J V, Nixon J H, Billingham J, and Richardson I M,* 'Review of the Technology of Deepwater Repairs for Offshore Structures', BOSS conference, Delft, 1997.

14. DESIGN RESTRICTIONS

The structural design of components is dependent on material properties, to ensure that there is an adequate margin of safety. This margin is usually reflected in allowable stresses being a proportion of the yield or ultimate stress, to ensure that tensile stresses are confined to the elastic region of the stress-strain curve. The design of some offshore components is more complex, e.g. tubular joints, where current design formulae include the UTS as a parameter. However most offshore codes also include a restriction on higher strength steels, limiting the design value of either the yield or ultimate stress, through the yield ratio (ratio of yield to ultimate strength). In addition, current codes and standards recommend a limit of Re/10 for Charpy toughness – but associated with a low temperature of -60°C (see section 6). The design of individual components will be assessed in the rest of this report.

14.1 BUCKLING OF MEMBERS

In terms of material properties, member buckling strength is governed by yield stress. The data which have been used to generate the design formulae are for medium strength steels and, for example, the draft ISO standard [14.01] includes the warning that application of the recommendations to higher strength steels (>500MPa) may lead to unconservative results. The ISO draft standard also provides a recommendation that the maximum value of yield to ultimate strength should not exceed 0.85; also the strain at the ultimate tensile strength should be at least 20 times the strain at the onset of yield. Many high strength steels have yield ratios exceeding 0.85 and the strain requirement may also be difficult to meet for these steels.

14.2 STATIC CAPACITY OF TUBULAR JOINTS

As already noted in section 3.3 and Chapter 4, the static capacity is proportional to the material UTS. The capacity equations have been developed from test data, almost all of which are from joints fabricated from steels with yield strengths <500MPa. Historically there has been a concern that the use of these equations to determine the strengths of joints with chords with higher yield strengths may result in lower post yield reserve strengths, which is the basis for joint design. There is therefore a requirement that the yield stress of the chord (in calculating the ultimate joint capacity) should not exceed a factor times the tensile strength of the chord for materials with a specified minimum yield strength of 500MPa or less.

As noted in section 3.3 and Chapter 4, this factor varies from 0.67 for the API code [14.02] to 0.7 for HSE Guidance [14.03]. Hence, as noted in section 3.3, current codes and standards which are used for the design of high strength steel joints do limit the benefit of using these steels in practice. The draft ISO standard [14.01] is proposing increasing this limit to 0.8, which could benefit the use of high strength steels offshore, although the draft standard also recommends that demonstration of adequate toughness is also a requirement for this limit to be extended. For tension loading there is some suggestion that the limit of 0.8 is unconservative [14.04] (based on very limited results) and that a lower limit of yield ratio may be appropriate (see section 3 and chapter 4).

14.3 DRAFT ISO STANDARD RECOMMENDATIONS FOR HIGH STRENGTH STEELS

As noted in section 4, a draft ISO standard [14.01] is being finalised which is expected to provide the way forward in future offshore standards. Table 14.1 lists the approach being taken in this towards the use of high strength steels. It can be seen that for steels with yield strengths over 500MPa most of the equations in the draft standard have limited applicability and in many cases test data are required to justify the use of high strength steels, with a consequent increase in cost and time. These limitations, necessary in view of the limited data available for high strength steels, are likely to continue to hinder the more widespread use of high strength steels offshore.

REFERENCES

14.01 ISO 'Petroleum & Natural Gas Industries, Fixed Steel Offshore Structures', ISO 19902, to be published

14.02 API 'Recommended Practice for Planning, Designing & constructing Fixed Offshore Platforms', 20th edition, 1993

14.03 HSE/D.Energy, 'Offshore Installations Guidance on Design, Construction & Certification', 1990, HMSO, London

14.04 *Nichols N W*, private communication

Table 14.1 - Approach in ISO 19902 to medium & high strength steels

Component/	Material restriction	Design limitation
Member - buckling	Yield strength less than or equal to 500MPa	Yield strength used in capacity equation limited by yield to ultimate ratio ≤ 0.85
		Strain at the ultimate tensile strength should be at least 20 times the strain at the onset of yield
	Yield strength > 500 MPa	No design recommendations
Joint - ultimate strength	Yield strength less than or equal to 500MPa	Yield stress of the chord (in calculating the ultimate joint capacity) should not exceed 0.8 times the tensile strength of the chord
		Yield strength used in equations limited by a yield to ultimate ratio ≤ 0.85, for compressive loading
	Yield strength > 500MPa	Capacity equations for compressive loading can be used with a limiting yield to ultimate ratio of 0.8, but dependent on material having demonstrated adequate ductility in HAZ & parent plate.
Joint - fatigue life	Yield strength less than 500MPa	Equations provided to calculate fatigue life
	Yield strength >500MPa	Equations provided are not applicable, test data or fracture mechanics to be used.
Corrosion Protection	Steels with specified min. yield strength >720MPa	Special considerations required for cathodic protection. Any welding or fabrication shall be carried out to qualified procedures which limits hardness to HV350.

15. SUMMARY AND CONCLUSIONS

Use of High Strength Steels

- There is an increasing use of HSS offshore, particularly in topside applications, but increasingly also in structures. Drilling jack-ups have been in use for many years, with satisfactory performance. Production jack-ups are a relatively new application with demanding requirements.

Mechanical Properties & Weldability

- As a result of recent developments, steels are now available with good combinations of high strength and toughness. However each steel grade has a range of yield strengths, a factor which is not fully recognised in design. There is a better understanding of welding requirements and welding of 450 grade steels is now routine. More work is required on improving the welding of higher strength steels and high strength welding consumables. On weld mismatch, there is an increasing problem with steels with YS > 600MPa, with a limited availability of suitable high strength weld metals to achieve the required degree of overmatching.

Codes & Standards

- In general current codes and standards do not cover steels with yield strengths greater than 500MPa, which limits their use offshore.
- Yield ratio is recognised as important but is a limited single measure of a complex process, concerning the plastic behaviour of HSS. Most high strength steels (σ_y > 500MPa) have yield ratios close to or greater than 0.85, which is in excess of the current design limits in codes and standards of 0.67 or 0.7 that apply to tubular joints. It is likely that these current limits will be relaxed to 0.8, particularly for lower strength steels (<500MPa), but some concerns remain about the treatment of YR for higher strength steels and particularly tubular joints in tension. A better understanding of this parameter or the development of an alternative measure is required.

Fracture

- Most current high strength steels have adequate toughness to give satisfactory performance offshore in both parent material and HAZ regions. Some problems of consistency in toughness can arise with very high strength weld metals.
- Some concerns have been raised about the sole use of Charpy data (from tests at a given temperature) for assurance of toughness in HSS. The measure provided in current codes and standards, based on Charpy values being not less than Re/10, may be too simplistic, and a fracture mechanics approach would be preferred. A correlation between such data and fracture mechanics parameters is needed for currently available steels.

Fatigue

- The fatigue strength for a given life does not improve in direct proportion to yield strength. Limited data indicate that some HSS have a fatigue life performance in seawater comparable to medium strength structural steels, under normal cathodic protection conditions. The much larger amount of air data shows a generally good fatigue performance, comparable to that of conventional medium strength steels. Crack propagation rates are also similar to medium strength structural steels, at a similar stress range.
- Some concerns still remain on the fatigue performance of HSS in seawater at high negative cathodic protection potentials when significant amounts of hydrogen are generated based on very limited data. Poor fatigue performance also results when hydrogen sulphide is present, even in small amounts, but there are limited data to quantify this.
- More data are still required for confident predictions of fatigue performance under a range of practical CP conditions; in the meanwhile producing test data for candidate high strength steels would appear to be the best approach.
- Weld improvement techniques offer significantly enhanced fatigue performance by delaying fatigue crack initiation for HSS. Some limited data show significant enhancement of properties

but there are cost implications and concerns about proving the technique in practice. Further work is required to enable this benefit to be used more widely in practice.

Cathodic Protection & Hydrogen Embrittlement

- Due to the increased sensitivity of HSS to hydrogen compared to medium strength steels more positive CP potentials are required in practice to ensure good field performance. Typical recommended values are more positive than –830Mv (Ag/AgCl).
- The use of voltage limiting diodes, as recommended by HSE, has shown some problems in practice and further work is required to establish clear design procedures.
- The availability of more positive potential anodes (e.g. those including gallium) may provide a better solution but field evaluation is required.
- Some codes and standards require protection potentials to be controlled in a narrow band (-770 to –830mV(Ag/AgCl)) which is extremely difficult to achieve in practice.
- The recommended technique for the testing of steels for hydrogen embrittlement (slow strain rate testing) is relatively quick to use but poorly defined, providing only a limited guide to performance in practice. The benchmarking of HSS against the performance of medium strength steels in seawater with CP (using slow strain rate testing) has limited value. A better defined test procedure is required for this important requirement of selecting suitable steels for minimal hydrogen sensitivity, probably based on fracture mechanics.

High Temperature Properties

- There is very little data on the high temperature properties of the type of high strength steels used offshore. Limited data at intermediate strength levels (450MPa) have indicated comparable performance to that specified in BS5950 Part 8. However, more data are required in order to give greater confidence in materials selection and, at present, test data are required to demonstrate satisfactory performance.

High Strain Rate Properties

- Offshore strain rates can vary over a very wide range from wave loading ($10^{-4}s^{-1}$) to $10^{+}6\ s^{-1}$ for blast effects. Fracture toughness, yield stress, UTS and hence yield ratio vary with strain rate. Unfortunately very little data exist for high strength steels (>450MPa). However, the limited data available indicates that mechanical properties can vary significantly with increasing strain rate and this needs to be taken into account in safety assessments, using test data where necessary.

Field performance

- Drilling jack-ups have extensive service experience, with generally good performance, but rely on regular dry dock inspection. Many of these older designs use HSS with poorer properties than those available with current HSS. The problem of hydrogen embrittlement is being addressed by more careful use of CP, use of biocides and in some cases by specifically using voltage limiting diodes to limit the range of negative potentials from CP. However cracking has been found in these jack-ups, even when the recommended measures for control of HAC have been implemented. The causes remain unclear as in most cases corrosion potentials in the field have not been measured. Hence, the success of these measures in practice still has to be proven.
- Production jack-ups (without the opportunity for dry dock inspection) have been deployed only since 1996, and hence have limited field experience to date. In these applications careful design has ensured HSS are not used in sensitive regions for hydrogen damage, e.g. close to the seabed. A range of CP options has been or is being used (voltage limiting diodes, coatings, conventional CP system); experience will provide guidance on their performance in practice. The design of these jack-ups is such that topside loading ensures mainly compressive loadings in the legs, and at critical nodes, with benefits to fatigue life.

Design Restrictions

- As already noted, most current codes and standards do not cover HSS with yield strengths >500MPa. The draft ISO standard which has been prepared over the last few years does not

improve on this position. The DNV code, however, does provide limited guidance for such steels. The use of HSS in practice therefore generally requires the provision of test data, which is costly and time consuming.

- The current restriction on yield ratio in codes and standards for tubular joints (0.67 – 0.7) does limit the use of HSS in practice. However, this limit is likely to be increased to 0.8 for HSS in tubular joints in compression.
- The management of high strength steels in seawater with CP is a key issue which needs to be addressed by further test evaluations and the establishment of preferred design parameters.
- In-service underwater inspection for hydrogen cracking needs further consideration as there is very limited experience to date. Repair methods for high strength steels under water may also need further development..

APPENDIX 1
OTHER STRUCTURAL APPLICATIONS OF HIGH STRENGTH STEELS - BOLTS AND THREADED FASTENERS

BOLTS & THREADED FASTENERS

In some offshore structural components (e.g. flanges and repair clamps), threaded fasteners are the primary means of transferring loads across the connector. Materials for such fasteners are usually selected on the basis of yield and ultimate strength, ductility and performance in seawater (e.g. corrosion resistance etc.), resistance to embrittlement, loss of stress due to relaxation and creep. Bolts and fasteners are normally pre-tensioned to a given level to provide resistance to disengagement and to increase fatigue performance.

Typical fastener/bolt materials are Grade 8.8, B7 and L7 steels, with yield strengths in the range 630-800MPa. The draft ISO standard [A1.01] recommends that the yield strength should be limited to 725MPa to avoid potential stress corrosion cracking. In addition the standard recommends that the level of cathodic protection should be -900mV +/- 50mV(Ag/AgCl).

Fatigue of fasteners is a significant design requirement. Until recently most fatigue data for such components was based on air testing. Recent tests have established new criteria, including the performance in seawater.

The British Standard 7608 [A1.02] includes an S-N curve for bolts in air which is presented in the form of log (N) versus constant - log (S/UTS), where S is the stress range and UTS the ultimate strength of the steel. It is found that this curve is conservative provided the UTS is limited to a maximum value of 785MPa. The recommended S-N Curve in the draft ISO standard [A1.01] for air is:

$$\text{Log } N = 11.55 - 3\log Sr$$

where Sr is the stress range in the bolt or fastener.

A thickness effect has also been found and is presented as a factor on stress, which is :

$$F = (d/20)^{0.3}$$, where d is the diameter of the fastener in mm.

A difference in fatigue performance has been found for bolts with threads which are cut or rolled. If rolling is carried out before heat treatment then the fatigue performance is similar to that observed for cut threads. If the heat treatment is after the rolling then improved life was observed.

For fatigue performance in a seawater environment with cathodic protection recent tests [A1.03] have shown that bolts of grades 8.8, B7 and L7 show a significant loss of fatigue life in seawater, relative to air performance. The draft ISO standard recommends a reduction factor of 3 on life for adequate cathodic protection in the range -850 to -1000mV (Ag/AgCl). For higher grade steels a recommendation is made that both the fatigue performance and the hydrogen embrittlement sensitivity should be addressed through a validated test programme.

The above concerns expressed in the draft ISO standard [A1.01] arise from some additional tests which have been undertaken on grade 12.9 bolts with a specified minimum yield strength of 1060MPa [A1.03]. Results from seawater testing of these bolts showed a more significant reduction in life compared to air performance, with reduction factors of about 5.4 for freely corroding conditions and 4.5 for cathodic protection (at -850mV).

REFERENCES

1.A1 ISO ' Petroleum & Natural Gas Industries, 'Fixed Steel Offshore Structures', ISO CD19902, to be published

1.A2 BS 7608 'Fatigue Design and Assessment of Steel Structures', BSI, London, 1993

1.A3 MaTSU 'Fatigue Performance of Threaded Connections in air and seawater environments - derivation of S-N curves', MATR 0425 report, 1997.

APPENDIX 3

Design Rules
If $\sigma_y < 0.67 \bullet \sigma_{UTS}$ – design capacity = σ_y

If $\sigma_y > 0.67 \bullet \sigma_{UTS}$ – design capacity = 0.67 σ_{UTS}

Case 1
Compare Steel A σ_y = 355MPa σ_{UTS} = 540MPa YR = 0.66
Design capacity = σ_y = 355

with Steel B σ_y = 550MPa (1.5 x σ_y of Steel A) σ_{UTS} = 610MPa YR = 0.9
Design capacity = 0.67σ_{UTS} = 0.67 x 610 = 403, i.e. only 13% increase in capacity when yield strength has increased by 50%

Case 2
Compare Steel C σ_y = 450MPa σ_{UTS} = 562MPa YR = 0.8
Design capacity = 0.67 x 562 = 375

with Steel D σ_y = 675MPa (1.5 x Steel C) σ_{UTS} = 733MPa YR = 0.92
Design capacity = 0.67 x 733 = 489, i.e. design capacity has increased by 30% when yield strength has increased by 50%.

Case 3
Compare Steel E σ_y = 355MPa σ_{UTS} = 540MPa YR = 0.66
Design capacity = 355

with Steel F σ_y = 390MPa (i.e. 10% increase in yield strength) σ_{UTS} = 520MPa
YR = 0.75
Design capacity = 0.67 x 520 = 348, i.e. design capacity of steel F is 2.5% lower than that of steel E even though the yield strength is 10% higher.

APPENDIX 6
FRACTURE TOUGHNESS CONCEPTS

A6.1 DUCTILE TO BRITTLE TRANSITION : AN INTRODUCTION TO FRACTURE

Plastic deformation in part of a structure represents a permanent change of shape and is a method of absorbing energy. All phases of steel except austenite are capable of undergoing a ductile-to-brittle transition as the temperature is reduced because of the increasing difficulty in undergoing plastic deformation. In structures, it is the behaviour of the steel in the presence of a notch, crack, stress concentrator, etc. that is important.

The Charpy V-notch impact energy represents the resistance of the material to impact loading in the presence of a standardised stress concentration, by recording the energy absorbed (Joules) from the pendulum by the specimen [6.A1]. Charpy results, however, cannot be considered to be directly relevant to structural behaviour. **This is because the extent of plastic deformation can depend on the geometry and the strain rate as well as the temperature.** The geometric effect is not related to the extra mechanical deformation of rolling required to reduce the thickness of the plate or section (although this plays a role). It is related to the local three-dimensional stressing, normally termed the 'stress state', for which the extremes are 'plane stress' and 'plane strain'.

Fracture mechanics tests (K tests, crack tip opening displacement and J-integral tests) have the advantage that they can use the full structural thickness to determine the toughness. The fracture toughness value that is calculated from these tests [6.A2; 6.A3] is quite different, in both concept and in terms of units ($N.m^{-3/2}$), from the Charpy impact toughness. **The fracture toughness value of ferritic steel also exhibits a ductile-to-brittle transition as the test temperature is reduced but the transition temperature for fracture mechanics tests of different geometries will be different from each other and different from the Charpy transition curve.** The transition temperature range may also differ.

Where crack propagation involves plastic deformation at the crack tip region, the energy of deformation has to be repeated at each stage of crack advance. However, if a brittle cleavage crack begins to propagate, it is not easy to stop it and relatively little deformation energy is expended even in large cross-sections. The problem is likely to be compounded because the larger sections will tend to generate plane strain conditions and restrict plastic deformation, i.e. give 'constrained yielding', because of the geometry.

Full-scale tests such as wide plate tests use the correct thickness and may retain more residual stress but may not exhibit the correct geometry. In addition, the choice of crack size and shape can effectively determine the amount of plasticity and whether or not a brittle fracture occurs. The tests, therefore, are useful for assessing the performance of a known flaw but are relatively expensive ways of determining a toughness value. The correlation between wide plate and smaller-scale standardised fracture mechanics tests is well established and the corrections required for crack geometry changes are backed up by calculations using finite element methods.

A6.2 CHARPY V-NOTCH VALUES : BACKGROUND

Charpy tests give a comparison between different materials using a relatively cheap standardised test that can be applied to different steel compositions, different batches, and different regions across a welded joint. An 'upper shelf' value is obtained when the failure occurs by only a ductile process, and improving the upper shelf Charpy energy improves the resistance of a structure to a running ductile failure. Reducing the transition temperature usually, but not always, implies an improved

margin of reserve against brittle fracture of a structure. The uncertainty arises because the effects of geometry and strain rate vary with material.

The specification of Charpy impact values in offshore applications relates more to proven past experience with comparable steels and weld metals than to any design-related criteria.

In the early days of offshore construction, a common specification for Charpy values was '20ft-lbs (27J) at -30°C and 30ft-lbs at -20°C'. It was quickly recognised that the offshore industry was using thicker steel plates as production moved to deeper water and this introduced the possibility of significant **residual stresses** from welds, which could influence the **stress state** in the structure.

Since **Charpy values do not reflect this increased risk of fracture**, the 1982 Code of Practice for Fixed Offshore Structures (BS 6235:1982) [6.A4] gave guidance on the Charpy test temperature at which an average of 27J should be recorded, according to plate thickness and design temperature [Figure A6.1]. The Department of Energy's document, 'Offshore installations: Guidance on design and construction' [6.A5], used the same approach but simplified the diagram by specifying a 27J requirement for weld metal and Grade 43 steel and a 34J (25ft-lbs) requirement for Grade 50 steel. The 1990 version [6.A6] of these guidance notes required 36J from 355MPa steels. For steels up to 100mm thick, the test temperature at which the specified energy should be achieved varied from -20°C to -40°C depending on material thickness, location, welded state and whether stress-relieved. For steels above 100mm thick, it was recommended that the test temperature be agreed with the certifying authorities.

Essentially, this procedure uses a temperature displacement for the onset of fracture in 10mm Charpy specimens to compensate for the constraint on generating plasticity in thicker sections subjected to a triaxial state of stress.

Impact toughness requirements for steel with SMYS above 355MPa were by agreement between parties. The corresponding requirements for welded steel bridges (BS 5400:1982) placed an upper limit on steel thickness according to strength and grade [6.A7]. Alternatively, enhanced Charpy requirements for greater thickness (t mm) could be sought according to

$$C_V \geq \frac{R_e}{355}\left(\frac{t}{2}\right)$$

where R_e is the specified minimum yield stress (SMYS) of the steel in MPa and the impact energy C_V is in Joules. The 355 value is the SMYS for the traditional Grades 50D and 50E steel, permitting adjustment for other BS 4360 steel grades.

The ratio of yield stresses in the above equation reflects the fact that a high strength steel is likely to carry a proportionally higher load and therefore requires a proportionally higher energy absorbing capability.

Norwegian 'Rules for the Design, Construction and Inspection of Offshore Structures' (DnV 1977, corrected 1981) dealt with the interaction between thickness, yield strength and toughness in a slightly different way [6.A8]. Like most of the offshore procedures, it dealt with thickness effects by means of an offset in the Charpy test temperature, rather than scaling the required value C_V with a thickness-related factor such as (t/2).

Impact toughness requirements for base metal, heat affected zone (HAZ) and weld metal were the same. A temperature offset for testing was specified in relation to the design temperature, and the required Charpy energy level was linked to the SMYS of the steel. 27J was required from longitudinal specimens for steels with yield stress up to 275MPa, then the requirements rose linearly to 39J for 390MPa steel. There was also a minimum toughness requirement for transverse specimens, rising from 18J to 26J [6.A8]. [Figure A6.2] This requirement relates to the fact that

inclusions, especially manganese sulphides and silicates, can become elongated as a result of rolling. In a transverse specimen, the crack propagates in a longitudinal direction and so the weakly-bonded inclusions can offer a low-energy path for part of the crack.

For simplicity, the above requirements are now usually expressed as **'R_e /10'**, so that the required longitudinal impact toughness in Joules is equal to one tenth of the SMYS in MPa, and the required transverse value is taken as two thirds of this. This appears to relate directly to the experience that 20ft-lbs (27J) at -30°C had been adequate for grade 43 steels with SMYS of 275MPa and the design assumption that higher strength steels carry a proportionally higher load.

A6.3 CHARPY V-NOTCH VALUES FOR HIGH STRENGTH STEEL

Early specifications for the offshore industry related only to steel with a SMYS of 355MPa that was produced by the normalisation route. The metallurgy therefore comprised a fine ferritic matrix with some pearlite consistent with a low C, leanly alloyed C-Mn steel. Further tempering as a result of welding or post weld heat treatment was likely to have relatively little effect on the ductile-to-brittle Charpy transition. However, care was required for the weld metal and heat affected zone regions that had been sufficiently heated to undergo a phase change as a result of welding, as these parts could produce the inherently brittle martensite structure if cooled sufficiently quickly.

The Charpy test was essentially a verification of an acceptable microstructure in these normalised steels. The specification increased the probability of initiating a cleavage fracture by Charpy testing at a low temperature. If the product avoided brittleness (by absorbing at least 27J of energy for example), the implication was that those microstructures that were most at risk of undergoing fracture because of triaxial loading and geometric constraint were absent.

The yield strength of steel can be enhanced further by thermo-mechanically controlled rolling and processing (TMCR and TMCP) to produce ultra-fine ferritic structures without the need for the reheat cycle of normalisation; but there is a limitation on the thickness of the product due to the requirements for rolling deformation. Products with SMYS around 450MPa were produced specifically for offshore structural use and generally exhibited excellent impact toughness, both in plate and HAZ, and low transition temperatures (frequently below -80°C). The weldability was also excellent. Some concerns were expressed about HAZ softening from grain growth but this appeared to be counter-balanced by local strain hardening and no marked loss in impact toughness.

Higher strength steel is usually manufactured using the quench-and-temper (Q & T) route. This produces a fundamentally different microstructure from the normalisation route. The quench produces a finely-structured martensite or bainite product that is usually stronger than required and is too brittle for direct use. The tempering reduces the strength, relieves some of the residual stress and improves the notch toughness. The repeatability of the quenching and the control of the tempering have important influences on the mechanical properties from plate to plate. As a consequence, when examining the impact test results from a single batch of steel, the transition temperature range tends to be wider and there can be appreciable scatter in recorded impact energy values in this range for some high strength steels.

An estimate of the linear elastic toughness (i.e. the plane strain fracture toughness K_{IC}) made from correlations with Charpy V-notch impact test data from the appropriate material microstructure is permitted in current BSI defect assessment methodology [6.A9]. However, it is pointed out that the majority of correlations relate only to ferritic steel plates, and that caution is required if using mean value or tolerance curve correlations rather than lower bound relationships. The first of two relationships in the document is intended as a lower-bound to known experimental results for determining K_{IC} from the temperature displacement from a selected Charpy impact energy. It is based on lower-bound data for ASTM A533 Type B steel. Data on other steels and weld metals have been shown to fall above this lower-bound curve, with the exception of some thick section tests on pressure

vessel steels. The second relationship imposes an upper limit to the toughness obtained by the correlation, and it is known that this may not always be conservative for assessing flaws in heat affected zone regions of BS4360 steels.

The implication therefore remains that it is preferable to determine directly the fracture toughness of the appropriate material for parent steel and weld regions when utilising higher strength steel.

A6.4 FRACTURE MECHANICS TESTS : BACKGROUND

A6.4.1 Brittle Materials

Fracture mechanics has evolved from studies of cracking in inherently brittle material. A material containing a crack under tensile loading, which is usually the most severe mode, experiences a rising applied stress intensity factor K_{app} as the load increases. In the simple geometry of an infinite plate containing a centre crack of length '2a' that goes all the way through the plate thickness, a first-order approximation to it is given by

$$K_{app} = \sigma_{app}\sqrt{(\pi.\bar{a}_{eff})}$$

where σ_{app} is the stress generated by the load in the absence of the crack, and $\bar{a}_{eff}$ is the 'effective crack length parameter'. For elastic failure (e.g. glass at low temperature), $\bar{a}_{eff}$ is exactly the same as the half-crack length 'a'. Most engineering materials exhibit some plastic deformation in the crack tip region before failure - good toughness being a selection criterion to promote structural safety - and $\bar{a}_{eff}$ is then somewhat larger than the measured value of 'a' to correct for the plasticity. Failure is expected when K_{app} reaches the fracture toughness of the material. In reality, reserve factors are required to improve confidence in the safety, hence failure is usually conceded at lower applied stress intensity levels [6.A9].

The (apparent) fracture toughness K_C is not a constant. Its value depends on stress state (geometry), strain rate and temperature. **The plane strain fracture toughness K_{IC}** (for tensile 'mode I' loading) depends only on strain rate and temperature: it is determined in practice when plane strain dominates the stress state and is regarded as the lower bound for a given temperature and strain rate. K_{IC} is sometimes called the material's toughness in an attempt to distinguish it from the higher apparent fracture toughness values.

Although the intended structural loading may be just simple uniaxial tension, the redistribution of stress in the vicinity of a stress concentration will result locally in stresses in at least two directions and for thick sections or complex geometries a significant amount of triaxial stressing can be produced. The surface layers will always be in plane stress but a crack in a thick section may experience plane strain conditions in the central regions.

In the absence of plasticity, the difference in toughness caused by a change in the elastic stress state between plane stress and plane strain is relatively small (around 10% and relates to $(1-\nu^2)$ where ν = Poisson's ratio). However, plastic deformation and strain hardening capabilities are desirable protection against overload for engineering structures.

When plasticity is expected, a move towards plane strain can significantly delay the onset of plasticity until higher applied loads and reduce the spread of plasticity in the crack tip region. The apparent toughness K_C from a small-scale test then easily can be more than twice the K_{IC} value [Figure A6.3]. It is therefore important that the fracture toughness value chosen to characterise the structure, or a specific region of it, is obtained for a representative stress state.

K test methodology recommends the use of standardised specimens where the geometry ensures that the thickness dimension is the principal factor controlling the state of stress. The geometry employs deep straight-fronted cracks in order to generate a high level of constraint on any plastic deformation, in an attempt to generate the minimum value of the fracture toughness for that particular thickness. The calculation of fracture toughness is based on elasticity theory corrected for small amounts of plastic deformation. For more ductile materials, similar geometry can be used but the calculations are based on elasto-plastic theory.

A6.4.2 Ductile materials

As crack tip plasticity increases, the error in calculating toughness using $\bar{a}_{eff}$ corrections to elastic equations for critical K_{app} values increases and therefore elasto-plastic parameters are preferred. The most common are J-integral ('J') and crack tip opening displacement (CTOD or δ) which are used to characterise crack initiation (i.e. when the crack starts to move). Resistance curves (R-curves) can be constructed from any fracture toughness parameter to deal with slow stable crack propagation but this concept is not usually applicable to offshore structures because of the dynamic loading.

J-integral has units of energy. The critical value J_C for a given stress state, strain rate and temperature is related to the apparent toughness but extends the range of calculation beyond that where elasticity-based K_C values are accurate, hence it is usual to denote critical toughness derived from 'J' tests as K_J. The fundamental relationship is

$$K_J^2 \approx E' J_C$$

where E' = E (Young's modulus) in plane stress and $E' = E/(1-\nu^2)$ in plane strain.

J_{IC} values, however, are at variance with the normal suffix coding and refer to the initiation of cracking caused by an instability in ductile micro-void coalescence (tearing) at the crack tip in mode I loading and do not imply that the values relate to plane strain conditions.

CTOD is a displacement between crack faces at the crack tip. Almost always it is a calculated value derived from crack mouth measurements. When the displacement results in a critical event at the crack tip, the CTOD value *characterises the fracture* and is *related to* the toughness but by itself it is not a fracture toughness value. If using CTOD, it is important to notice that it is the critical value of ($\sigma_{YP}.\delta$) that relates to fracture toughness K_C and not just the displacement δ_{CRIT}. It is usual to use δ_c δ_u and δ_i to signify 'fracture without prior tearing', 'fracture resulting from a slow-stable tear becoming unstable' and 'the point where ductile tearing initiated'. For a ductile material, these events may not be detected and the CTOD calculation may use the (first appearance of) maximum load value, which should be quoted as δ_m. This value depends on the geometry of the test specimen, is not linked to a fracture event and so does not characterise fracture. In the derivation of CTOD, it was assumed that crack tip plasticity was small scale, and under these conditions the relationship is

$$K_Q^2 = E'.m.\sigma_{YP}.\delta_{CRIT}$$

where K_Q is the toughness (written as K_C unless the conditions qualify as plane strain, in which case K_Q is K_{IC}), E' = E (Young's modulus) in plane stress and $E' = E/(1-\nu^2)$ in plane strain, m = a magnification factor taken as 1 in plane stress and 2 in plane strain, σ_{YP} is the uniaxial yield stress and δ_{CRIT} is the critical value of CTOD (which can be δ_c δ_u or δ_i). In practice, usually the stress state is not known, so there is an iterative loop to establish whether plane strain applied to the experimental value ... K_Q is evaluated assuming plane strain, and then a minimum dimension B_Q is evaluated from

$$B_Q = 2\tfrac{1}{2}.\left(\frac{K_Q}{\sigma_{YP}}\right)^2$$

If the actual specimen thickness B (and other specified measurements) are greater or equal to B_Q then the calculated toughness value K_Q can be quoted as K_{IC} rather than K_C.

The magnification factor m effectively gives the local constraint on yielding, ($m.\sigma_{YP}$) and its value is not readily determined. Flaw assessment procedures, therefore, should not attempt to compare K (and K_J) test values with CTOD quality requirements, or vice versa.

Wide Plate Tests are usually full-thickness samples about 1 metre square, but there is no standardised size, shape or method of production for the crack. The test is regarded as a full-size test that therefore is likely to be more representative of structural behaviour than the smaller fracture mechanics tests. This may be true in terms of residual stress fields for welded plates, and stress redistribution around a crack, but it is not always understood that the choice of crack geometry can have a significant influence on the result obtained. Wide plate tests therefore can be used to simulate behaviour where the crack geometry is known … although it may be necessary to move away from the simple plate geometry to something that better models the structure. However, it is important to realise that wide plate test results have been used extensively in the validation of correction factors for geometry and plasticity in defect assessment procedures. These procedures enable small-scale fracture tests to be used, and permit calculations of the effect of different levels of residual stress to be made [6.A9].

A6.5 FRACTURE MECHANICS VALUES FOR HIGH STRENGTH STEELS

In general, recent modern steel-making developments for structural steels that produce ultra-fine grained low alloy products have much-improved upper-shelf toughness and significantly lower transition temperatures for the same thickness compared with the older products of comparable strength.

To increase the strength of a weldable steel, the options are to reduce the grain size, to increase the alloy content and to produce an inherently stronger crystal structure (martensite, bainite). Refining the grain size is the best option, because toughness is improved as well as strength. Grain growth controllers are required in weldable steels, to prevent significant diffusional grain growth and hence loss of strength in the grain-coarsened HAZ regions. Q & T steels combine crystal structure change with fine grains. Fracture toughness tests, which relate to initiation of crack movement, are particularly likely to show scatter in the transition region when the microstructure is variable. Increasing the alloy content also tends to reduce toughness while increasing strength. A potential problem exists in these steels as a consequence of local precipitation changes in the HAZ regions of welds, also producing scatter in fracture toughness values. Regions causing this scatter have been called 'local brittle zones' or LBZs.

Another potential difficulty relates to the steepness of the fracture toughness transition from ductile to brittle behaviour in some modern high strength steels. As the size of the test-piece is increased, the constraint on yielding increases in the centre. The transition temperature moves towards higher temperatures but at the same time the upper shelf (apparent) toughness levels increase and the transition tends to occur over a narrower range of temperature (but usually with scatter). [See Figure 6.1]. Because of the variety of microstructures available for high strength steels and the increased levels of loading, it seems likely that the gradient of the transition curve may vary considerably among different steels and for different thicknesses. This may have important implications for the degree of safety (or degree of conservatism) when relying on toughness verification by Charpy impact testing of high strength steels. The customary values for the temperature offset (between the Charpy test temperature and the design temperature of the structure) may no longer be equally applicable to all high strength steels.

A6.6 FLAW ASSESSMENT CONSIDERATIONS FOR HIGH STRENGTH STEELS

Higher strength steels offer the possibility of weight reduction. Because of the thinner sections, there also may be cost benefits in terms of reduced welding. To maximise these gains when switching to higher strength steel, the design stress is increased by the same ratio as the yield strength increased. It is not immediately clear what requirement should be sought for the new toughness. There is also a further problem. Unless the resistance of the new material to mechanisms of crack growth, such as fatigue and stress corrosion cracking, also shows a comparable level of improvement, there now appears to be a disadvantage in having a reduced cross-section of steel. This is a result of maximising the structural stress level.

To exploit the potential benefits of high strength steels more fully, it may be necessary to establish a better combination in terms of stressing, resistance to fatigue etc and fracture toughness. Industry tends to utilise a design stress that is a particular fraction f of the specified minimum yield stress, such that $\sigma_{app} = f.\sigma_{YP}$. Moving to high strength steel at a slightly reduced value of f still permits higher applied stresses to be carried by the structure ... and the fatigue prognosis of the high strength steel has been improved as a result of the change in f.

The detail of such changes needs to be checked using recognised engineering critical assessment procedures such as BS 7910:1999 [6.A9]. However, the following section may be helpful in illustrating the approximate changes that can be expected from adjustments to some of the parameters.

If a high strength steel of SMYS $(\sigma_{YP})_2$ is being considered as an alternative to a more conventional steel of SMYS $(\sigma_{YP})_1$ then the associated changes in fracture toughness requirements can be estimated from the basic relationships. These can be related to the design stress expressed here as a fraction f of the yield stress. For clarity, plane stress conditions are shown, and the suffix letters indicating critical conditions are omitted (i.e. K_C is shown as K).

The basic relationships give $\dfrac{K^2}{E} \approx J \approx (\sigma_{YP}.\delta)$ and $\dfrac{K^2}{\pi.(f.\sigma_{YP})^2} = \bar{a}_{eff}$

Young's modulus E is normally assumed to be the same for engineering steels. However, the yield stress σ_{YP} varies with the steel, heat treatment etc. If using CTOD, it is important to notice that it is the critical value of $(\sigma_{YP}.\delta)$ that relates to fracture toughness K and not just the displacement δ. Selecting a higher yield strength material and specifying the same minimum value for the CTOD is equivalent to specifying a higher toughness. An improved toughness may be necessary if the stress will be increased, but that requirement is linked to the flaw size.

If the tolerable / critical flaw size is to remain constant, then a first estimate of the change in fracture toughness requirements may be obtained from

$$K_2 \approx \frac{f_2.(\sigma_{YP})_2}{f_1.(\sigma_{YP})_1} \, . \, K_1$$

$$J_2 \approx \left(\frac{f_2.(\sigma_{YP})_2}{f_1.(\sigma_{YP})_1}\right)^2 . \, J_1$$

$$\delta_2 \approx \left(\frac{f_2}{f_1}\right)^2 . \left(\frac{(\sigma_{YP})_2}{(\sigma_{YP})_1}\right) . \delta_1$$

However, some additional important factors must be considered relating to the stress state. It is likely that the designer will wish to increase the design stress in line with the SMYS of the material, hence f_2 / f_1 will be close to one, and if the section width is unaltered then the thickness in the structure will be reduced from B_1 to B_2 where

$$B_2 = \frac{f_1.(\sigma_{YP})_1}{f_2.(\sigma_{YP})_2} . B_1$$

The unwary may assume that the reduction in thickness B for the high strength steel will give a shift towards plane stress. However, the reverse is more likely because the thickness requirement for plane strain conditions to dominate is $B \geq B_Q$ where

$$B_Q = 2\tfrac{1}{2}.\left(\frac{K_{IC}}{\sigma_{YP}}\right)^2$$

The ratio B / B_Q is an indication of the state of stress. If the ratio is below one, it is K_C rather than K_{IC} that is applicable in a simple geometry where thickness controls the stress state, and this apparent toughness K_C depends on thickness [Figure A6.3].

If it is assumed again that a higher strength steel is used at constant width but in thinner section, then a shift towards plane strain is expected to occur if

$$\left(\frac{B_2}{(B_Q)_2}.\frac{(B_Q)_1}{B_1}\right) > 1 \qquad \text{i.e. if} \qquad \frac{f_1}{f_2}.\frac{(\sigma_{YP})_2}{(\sigma_{YP})_1}.\left(\frac{(K_{IC})_1}{(K_{IC})_2}\right)^2 > 1$$

It can be seen that the increase in yield stress $(\sigma_{YP})_2 > (\sigma_{YP})_1$ tends to increase this ratio and in general it is difficult to compensate for this by an increase in the material toughness K_{IC} at the same time as increasing the yield stress σ_{YP}.

This implies that the thinner sections of steel that can be considered for structural use when higher strength steels are selected do not also imply a move towards plane stress and its associated lower constraint on plastic deformation. It is possible that the reduced section will give a shift towards plane strain and the apparent toughness K_C measured for the chosen thickness will be correspondingly nearer the material toughness K_{IC} because of the constraint on plasticity.

The K_{IC} value is likely to be different for each steel but, because the constraint is also different, the effect on plastic deformation at the crack tip cannot be deduced reliably from either the ratio of the K_{IC} values or the ratio of K_C values at a common thickness. The K_C value appropriate to the thickness therefore should be used.

The relationships between fracture toughness parameters, as well as the calculation of values from tests, have to be corrected for this constraint, which will involve the $(1-\nu^2)$ term where ν = Poisson's ratio and the more significant adjustments to $\bar{a}_{eff}$ and the CTOD magnification factor m. Formal defect assessment procedures give guidance on these factors.

REFERENCES – APPENDIX 6

6.A1 BS 131:1971, 'Notched Bar Tests', Part 2, 'The Charpy V-notch impact test on metals', British Standards Institution, London. Now superseded by BS 131-6: 1988, 'Notched bar tests. Methods for precision determination of Charpy V-notch impact energies for metals', and by BS EN 10045-1: 1990, 'Charpy impact tests on metallic materials. Test method (V- and U-notches)'

6.A2 BS 7448: 1991, 'Fracture mechanics toughness tests' Part 1: 'Method for determination of K_{IC}, critical CTOD and critical J values of metallic materials', British Standards Institution, London

6.A3 BS 7448: 1997, 'Fracture mechanics toughness tests' Part 2: 'Method for determination of K_{IC}, critical CTOD and critical J values of welds in metallic materials', British Standards Institution, London

6.A4 BS 6235:1982, 'Code of Practice for Fixed Offshore Structures', British Standards Institution, London [Now Withdrawn]

6.A5 Department of Energy, 'Offshore installations: Guidance on design and construction', HMSO London, 3rd edition, 1984 [Now Withdrawn]

6.A6 Health and Safety Executive / Department of Energy, 'Offshore installations: Guidance on design, construction and certification', HMSO, London, 4th edition, 1990 [Now Withdrawn]

6.A7 BS 5400:1982, 'Steel, concrete and composite bridges. Code of practice for design of steel bridges', British Standards Institution, London

6.A8 Det Norske Veritas, 'Rules for the Design, Construction and Inspection of Offshore Structures', DnV, Høvik, Norway, 1977, corrected 1981

6.A9 BS 7910:1999 'Guide on methods for assessing the acceptability of flaws in metallic structures', British Standards Institution, London

6.A10 *Anderson T L*, 'Fracture Mechanics: Fundamentals and Applications', CRC Press, ISBN 0-8493-4277-5 (1991)

Figure A6.1 – 1982 Code guidance on Charpy V-notch requirements and steel grades for use offshore – (a) as welded construction, Grade 50 steel; (b) as welded construction, Grade 43 steel; (c) stress relieved construction [6.A4]

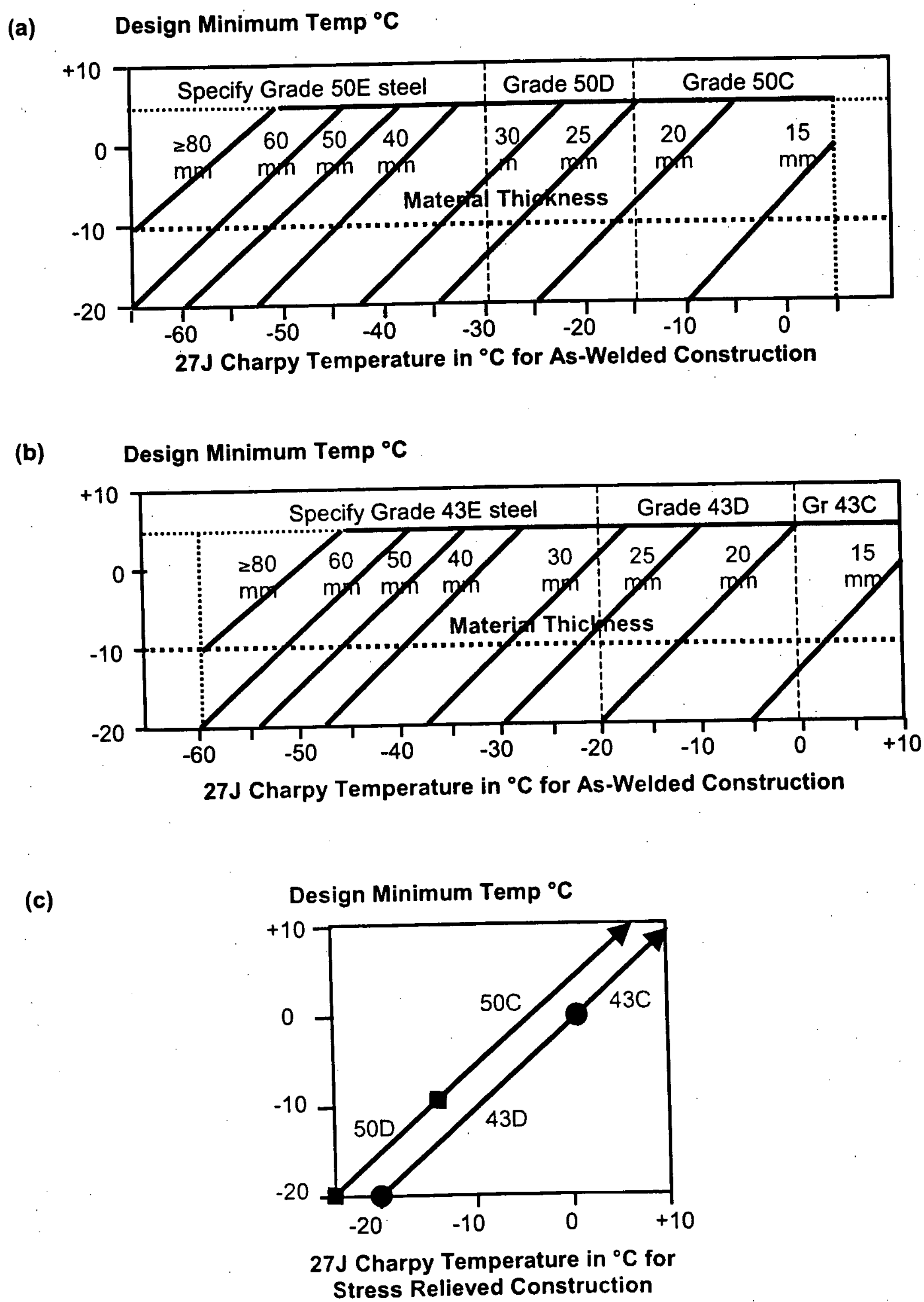


Figure A6.2 – Norwegian rules for the average minimum Charpy V-notch energy absorption for offshore structures and the associated test temperature [6.A8]

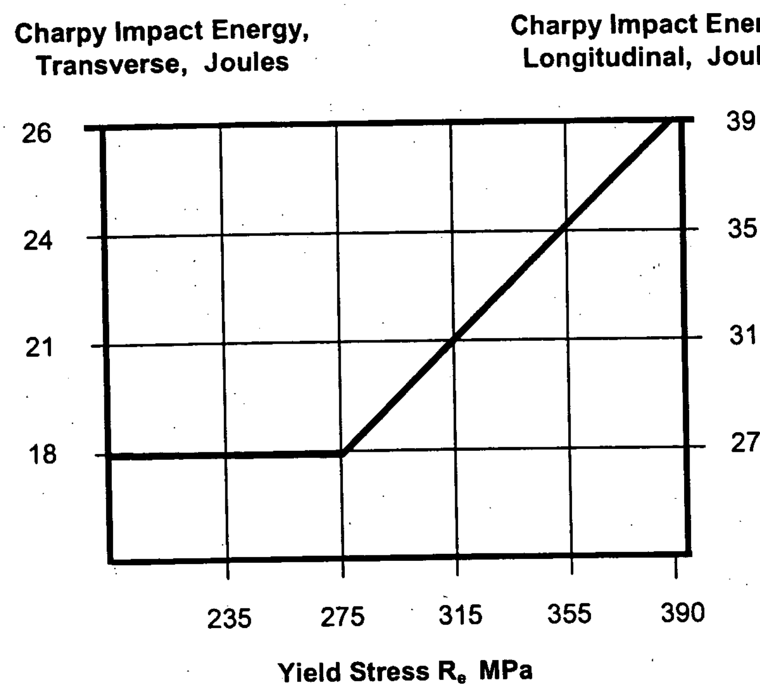
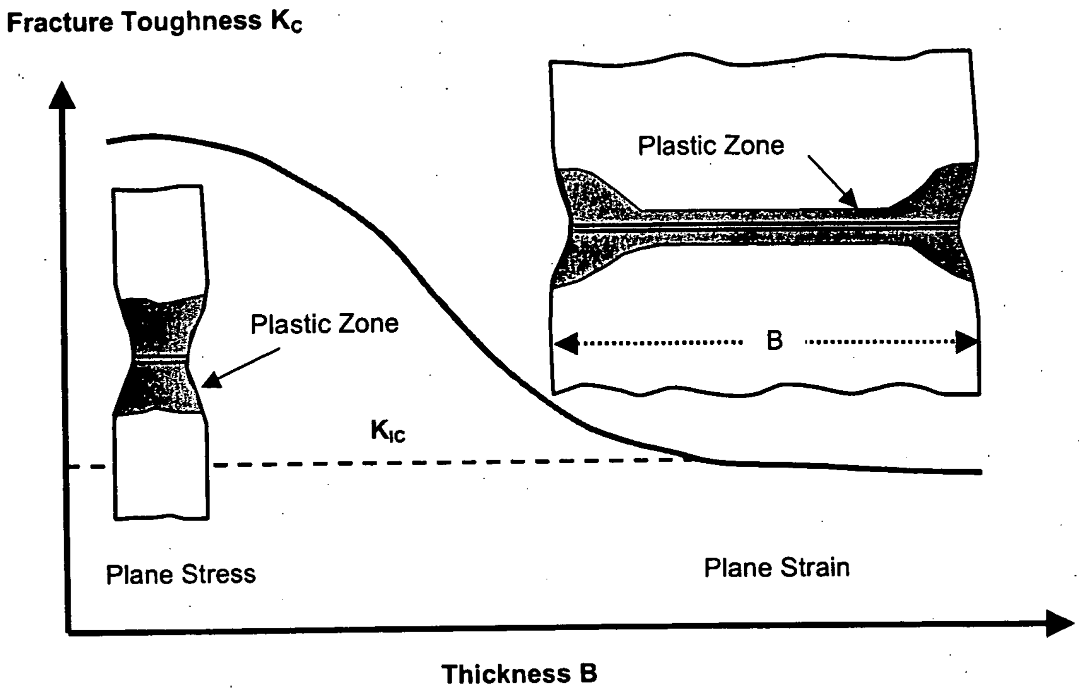


Figure A6.3 – Effect of specimen thickness on the mode I (opening mode) apparent fracture toughness K_C [6.A10]



APPENDIX 9

A9.1 SLOW STRAIN RATE TESTING

The HE susceptibility of steels in seawater is frequently assessed using SSRT as it is a relatively rapid, comparative method. However, it is important to appreciate the limitations of the technique as erroneous results can occur. For example, when welded specimens are tested the majority of the strain occurs preferentially in the softest part of the microstructure. This leads to failure in that region, whereas in practice HE usually occurs in the hardest region of the weld.

It is important that the steel undergoing SSRT has a bulk hydrogen concentration that is representative of the hydrogen level that exists in real structures after long-term exposure in service. As SSRT is a rapid method it is important for specimens to be precharged with hydrogen, prior to testing, by exposing them to the test environment for a sufficiently long period. Precharging times can be calculated from knowledge of the hydrogen diffusion coefficient in the material but unfortunately, literature values for diffusion coefficients [9.13] often vary by an order of magnitude. Predicted times required for precharging with hydrogen to produce a near uniform concentration profile are shown graphically for two values of diffusion coefficient in Figure A9.1. Precharging times would be significantly reduced if, for example, the calculation was based on achieving 50% of the surface concentration at the centre of the specimen. It should also be noted that hydrogen diffusion coefficients are very sensitive to temperature and are affected by stress.

Another area of concern with SSRT is related to replicating the effects of SRB on hydrogen pick-up. When dealing with sulphide problems in the past, H_2S saturated seawater solutions (3000ppm) have been used. SRB activity is unlikely to produce more than 600ppm and actual concentrations will probably be far lower. It also appears that whilst a number of workers have found that the effects of chemical sulphide differ from biogenic sulphide they disagree to which is the most severe. From a practical point of view the use of a simulated environment using chemical sulphide additions (or other chemical addition) is far more attractive than the dynamic SRB alternative due to the difficulty in maintaining constant bacterial activity and thus static sulphide levels.

BS4360 grade 50D steel has been used widely for offshore construction and is generally believed to have low susceptibility to HE, so it would seem reasonable to consider SSRT results for this material as a baseline for comparing other offshore steels. One critical parameter in SSRT is the strain rate used and this is often specified as $1x10^{-6}s^{-1}$ (cross head speed/gauge length). It has been found that embrittlement is greater when a slower rate is used but such rates are not used as they increase the test time. This does raise the concern that the faster rate may favour one material over the other when comparing two materials.

A9.2 FRACTURE MECHANICS TESTING

The fracture mechanics approach to embrittlement testing has the advantage that it can provide values for the threshold stress intensity factor (K_{th}) that must be applied to a material before crack propagation can occur. These values are especially useful as they can be employed directly for assessing the risk of a crack growing from existing flaws within a structural component. There is not a single K_{th} value for HE in a particular steel but a range of values which depend on the concentration of hydrogen absorbed from the environment. Values of K_{th} for high strength steels in a range on environments, including seawater and H_2S, have been reviewed by Gangloff [9.14].

Printed and published by the Health and Safety Executive
C0.6 09/03

MECHANICAL DESIGN AND SYSTEMS HANDBOOK

HAROLD A. ROTHBART, Editor-in-Chief

Dean, College of Science and Engineering
Fairleigh Dickinson University
Teaneck, New Jersey

McGRAW-HILL BOOK COMPANY

New York Toronto London

MECHANICAL DESIGN AND SYSTEMS HANDBOOK

 Library of Congress Catalog Card Number: 62-21118

54019

3 4 5 6 7 8 - MP - 1 0 9 8

CONTRIBUTORS

CLAYTON H. ALLEN, Ph.D., *Senior Engineering Scientist, Bolt Beranek and Newman, Inc., Cambridge, Mass.* (Section 7, *Noise Control*)

JOHN W. AXELSON, Ph.D., *Section Chief, Packings Development, Johns Manville Corp., Research Center, Manville, N.J.* (Section 24, *Pressure Components (Seals)*)

W. H. BAIER, Ph.D., *Senior Research Engineer, Armour Research Foundation, Chicago, Ill.* (Section 29, *Belts*)

ANTONIO F. BALDO, Ph.D., *Associate Professor of Mechanical Engineering, The City University of New York, New York, N.Y.* (Section 34, *Dampers*)

PHILIP BARKAN, Ph.D., *Senior Research Engineer, General Electric Company, Switchgear and Control Division, Philadelphia, Pa.* (Section 16, *Impact*)

STANLEY J. BECKER, M.S., *Engineer, Advanced Techniques Re-entry Systems Department, General Electric Corp., Philadelphia, Pa.* (Section 23, *Shrink- and Press-fitted Assemblies*)

KARL BRUNELL, B.M.E., *Research Associate, Princeton Laboratory, American Can Co., Princeton, N.J.* (Section 26, *Power Screws*)

JOHN A. CARLSON, Ph.D., *Engineering Specialist, MRD Division, General American Transportation Corp., Niles, Ill.* (Section 2, *Computers*)

SAUL FENSTER, Ph.D., *Chairman, Department of Mechanical Engineering, Fairleigh Dickinson University, Teaneck, N.J.* (Section 1, *Mathematics*, and Section 18, *Mechanical Design*)

E. T. FORTINI, M.S., *Staff Engineer, Compugraphic Corp., Reading, Mass., and Lecturer in Mechanical Engineering, Lincoln College, Northeastern University, Boston, Mass.* (Section 19, *Fabrication Principles*)

FERDINAND FREUDENSTEIN, Ph.D., *Professor of Mechanical Engineering, Columbia University, New York, N.Y.* (Section 4, *Kinematics of Mechanisms*)

THEODORE GELA, D.Sc., *Professor of Metallurgy, Stevens Institute of Technology, Hoboken, N.J.* (Section 17, *Properties of Engineering Materials*)

HERBERT H. GOULD, M.M.E., *Engineer, Sperry Gyroscope Co., Great Neck, N.Y.* (Section 1, *Mathematics* and Section 18, *Mechanical Design*)

14. Timoshenko, S., and J. N. Goodier: "Theory of Elasticity," 2d ed., McGraw-Hill Book Company, Inc., New York, 1951.
15. Donnell, L. H.: Longitudinal Wave Transmission and Impact, *Trans. ASME*, vol. 2, p. 153, 1930.
16. Sears, J. E.: On the Impact of Bars with Rounded Ends, *Trans. Cambridge Phil. Soc.*, vol. 21, 1909–1911.
17. Timoshenko, S.: Zur Frage nach der Wirkung eines Stosse auf einer Balken, *Z. Math. Physik*, vol. 62, p. 193.
18. Lee, E. H.: The Impact of a Mass Striking a Beam, *J. Appl. Mech.*, vol. 62, p. A-129, December, 1940.
19. Krefeld, W. J., M. G. Salvadori, et al.: An Impact Investigation of Beams with Butt-welded Splices under Impact, *Welding J. N.Y. Res. Suppl.*, vol. 12, no. 7, pp. 372S–432S, 1947.
20. Hoppmann, W. H.: Impact of a Mass on a Damped, Elastically Supported Beam, *J. Appl. Mech.*, vol. 70, p. 125, June, 1948.
21. Zener, C., and H. Feshbach: A Method of Calculating Energy Losses during Impact, *J. Appl. Mech.*, June, 1939, p. A67.
22. Clark, D. S.: The Behavior of Metals under Dynamic Loading, *Metal Prog.*, November, 1953, p. 67.
23. Brooks, W. A., and I. W. Wilder: Effect of Dynamic Loading on Strength of an Inelastic Column, *NACA Tech Note* 3077, March, 1954.
24. Hartz, B. J., and R. W. Clough: Inelastic Response of Columns to Dynamic Loading, *Proc. ASCE, J. Eng. Mech. Div.*, vol. 83, no. Em2, April, 1957.
25. Gerard, G., and H. Becker: Column Behavior under Conditions of Compressive Stress Wave Propagation, *J. Appl. Phys.*, vol. 22, no. 10, p. 1298, 1951.
26. Coppa, A. P.: On the Mechanism of Buckling of a Circular Cylindrical Shell under Longitudinal Impact, *Gen. Elec. Rept.* R60SD494, Missile & Space Vehicle Dept., Philadelphia.
27. Zener, C., and J. H. Hollomon: Plastic Flow and Rupture in Metals, *Trans. ASM*, vol. 33, pp. 188–226, 1944. Morkovin, D.: *ibid.*, p. 221 (discussion).
28. Davidenkov, N. N., and E. I. Belyaeva: Investigation of Fatigue under Repeated Impact, *Brutcher Trans.* no. 4719, POB 157, Altadena, Calif., Russian original in *Metalloved. i Obrabotka Metal.*, no. 11, November, 1956.
29. Davidenkov, N. N., and E. I. Belyaeva: Study of Impact Fatigue Strength, *Brutcher Trans.*, no. 4349, Russian original in *Metalloved. i Obrabotka Metal.*, no. 9, September, 1958.
30. Welch, W. P.: "ASME Handbook—Metals Engineering—Design," Chap. 8, Impact Considerations in Design, McGraw-Hill Book Company, Inc., New York, 1953.
31. Rayleigh, J. W. S.: On the Production of Vibrations by Forces of Relatively Long Duration with Application to the Theory of Collisions, *Phil. Mag.*, ser. 6, vol. 11, 1906.
32. Dohrwend, C. O., D. C. Drucker, and P. Moore: Transverse Impact Transients, *Proc. SESA*, vol. 1, no. 2, pp. 1–11, 1944.
33. Campbell, J. D., and J. Duby: The Yield Behavior of Mild Steel in Dynamic Compression, *Proc. Roy. Soc. (London)*, vol. A236, 1956.
34. Maiden, C. J., and J. D. Campbell: The Static and Dynamic Strength of a Carbon Steel at Low Temperatures, *Phil. Mag.*, ser. 8, vol. 3, 1958.
35. Johnson, J. E., D. S. Wood, and D. S. Clark: *J. Appl. Mech.*, December, 1953, p. 523.
36. Osgood, W. R. (ed.): "Residual Stresses in Metals and Metals Construction," chapter by J. Miklowitz, The Effect of Residual Stresses on High-speed Impact Resistance, Reinhold Publishing Corporation, New York, 1954.
37. Zener, C., and J. H. Hollomon: Effect of Strain Rate upon Plastic Flow in Steel, *J. Appl. Phys.*, vol. 15, 1944.
38. Whittaker, E. T.: "A Treatise on the Analytical Dynamics of Particles and Rigid Bodies," Cambridge University Press, New York, 1937.
39. Davidenkov, N. N.: Allowable Working Stresses under Impact, *Trans. ASME*, vol. 56, 1934.
40. Cardillo, R. M., and D. F. Kruse: "Load Bearing Characteristics of Butyl Rubber," ASME paper 61-WA-335.
41. Barkan, P., and M. F. Sirkin: "Impact Behavior of Elastomers," ASME annual mtg. paper, 1962; published in *Trans. Soc. of Plastics Engrs.*, July, 1963, pp. 210–219.
42. Bradley, J. N.: "Shock Waves in Chemistry and Physics," John Wiley & Sons, Inc., New York, 1962.
43. Rice, M. H., R. G. McQueen, and J. M. Walsh: "Compression of Solids by Strong Shock Waves," appears in "Solid State Physics," vol. 6 (Seitz and Turnbull, eds.), Academic Press Inc., New York, 1958.
44. Maxwell, B.: Mechanical Properties of Plastic Dielectrics, *Elec. Mfg.*, vol. 58, p. 146, September, 1956.

Section 17

PROPERTIES OF ENGINEERING MATERIALS

By

THEODORE GELA, D.Sc., *Professor of Metallurgy, Stevens Institute of Technology, Hoboken, N.J.*

CONTENTS

17.1. MATERIAL-SELECTION CRITERIA IN ENGINEERING DESIGN

The selection of materials for engineering components and devices depends upon a knowledge of material properties and behavior in particular environmental states. Although a criterion for the choice of material in critically designed parts relates to the performance in a field test, it is usual in preliminary design to use appropriate data obtained from standardized tests. The following considerations are important in material selection:

1. Elastic properties
   a. Stiffness and rigidity
2. Plastic properties
   a. Yield conditions, stress-strain relations, and hysteresis

3. Time-dependent properties
   a. Elastic phenomenon (damping capacity), creep, relaxation, and strain-rate effect
4. Fracture phenomenon
   a. Crack propagation, fatigue, and ductile to brittle transition
5. Thermal properties
   a. Thermal expansion, thermal conductivity, and specific heat
6. Chemical interactions with environment
   a. Oxidation, corrosion, and diffusion

It is good design practice to analyze the conditions under which test data were obtained and to use the data most pertinent to anticipated service conditions.

The challenge that an advancing technology imposes on the engineer, in specifying treatments to meet stringent material requirements, implies a need for a basic approach which relates properties to structure in metals. As a consequence of the mechanical, thermal, and metallurgical treatments of metals, it is advantageous to explore, for example, the nature of induced internal stresses as well as the processes of stress relief. Better material performance may ensue when particular treatments can be specified to alter the structure in metals so that the likelihood of premature failure in service is lessened. Some of the following concepts are both basic and important:

1. Lattice structure of metals
   a. Imperfections, anisotropy, and deformation mechanisms
2. Phase relations in alloys
   a. Equilibrium diagrams
3. Kinetic reactions in the solid state
   a. Heat-treatment by nucleation and by diffusionless processes, precipitation hardening, diffusion, and oxidation
4. Surface treatments
   a. Chemical and structural changes in carburizing, nitriding, and localized heating
5. Metallurgical bonds
   a. Welded and brazed joints

17.2. STRENGTH PROPERTIES—TENSILE TEST AT ROOM TEMPERATURE

The yield strength determined by a specified offset, 0.2 per cent strain, from a stress-strain diagram is an important and widely used property for the design of statically loaded members exhibiting elastic behavior. This property is derived from a test in which the following conditions are normally controlled: surface condition of standard specimen is specified; load is axial; the strain rate is low, i.e., about 10^{-3} in./in./sec; and grain size is known. Appropriate safety factors are applied to the yield strength to allow for uncertainties in the calculated stress and stresses in the calculated stress and for possible concentration factors and for possible overloads in service. Since relatively small safety factors are used in critically stressed aircraft materials, a proof stress at 0.01 per cent strain offset is used because this more nearly approaches the proportional limit for elastic behavior in the material. A typical stress-strain plot from a tensile test is shown in Fig. 17.1 indicating the elastic and plastic behaviors. In order to effect more meaningful comparisons in design strength properties among materials having different specific gravities, the strength property can be divided by the specific gravity, giving units of psi per pound per cubic inch.

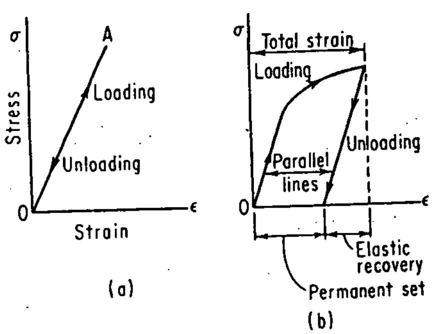


FIG. 17.1. Portions of tensile stress σ–strain E curves in metals.[1] (a) Elastic behavior. (b) Elastic and plastic behaviors.

The modulus of elasticity is a measure of the stiffness or rigidity in a material. Values of the modulus normally are not exactly determined quantities, and typical values are commonly reported for a given material. When a material is selected on the basis of a high modulus, the tendency toward whip and vibration in shaft or rod applications is reduced. These effects can lead to uneven wear. Furthermore the modulus assumes particular importance in the design of springs and diaphragms, which necessitate a definite degree of motion for a definite load. In this connection, selection of a higher-modulus material can lead to a thinner cross section.

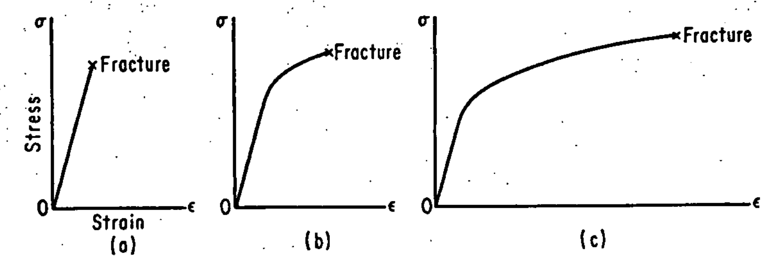


FIG. 17.2. The effects of treatments on tensile characteristics of a metal.[1] (a) Perfectly brittle (embrittled)—all elastic behavior. (b) Low ductility (hardened)—elastic plus plastic behaviors. (c) Ductile (softened)—elastic plus much plastic behaviors.

The ultimate tensile strength and the ductility, per cent elongation in inches per inch or per cent reduction in area at fracture are other properties frequently reported from tensile tests. These serve as qualitative measures reflecting the ability of a material in deforming plastically after being stressed beyond the elastic region. The strength properties and ductility of a material subjected to different treatments can vary widely. This is illustrated in Fig. 17.2. When the yield strength is raised by treatment to a high value, i.e., greater than two-thirds of the tensile strength, special concern should be given to the likelihood of tensile failures by small overloads in service. Members subjected solely to compressive stress may be made from high-yield-strength materials which result in weight reduction.

When failures are examined in statically loaded tensile specimens of circular section, they can exhibit a cup-and-cone fracture characteristic of a ductile material or on the other extreme a brittle fracture in which little or no necking down is apparent. Upon loading the specimen to the plastic region, axial, tangential, and radial stresses are induced. In a ductile material the initial crack forms in the center where the triaxial stresses become equally large, while at the surface the radial component is small and the deformation is principally by biaxial shear. On the other hand, an embrittled material exhibits no such tendency for shear and the fracture is normal to the loading axis. Some types of failures in round tensile specimens are shown in Fig. 17.3.

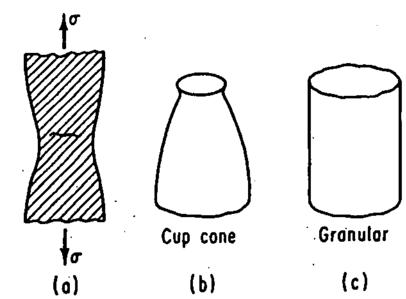


FIG. 17.3. Typical tensile-test fractures.[1] (a) Initial crack formation. (b) Ductile material. (c) Brittle material.

The properties of some wrought metals presented in Table 17.1 serve to show the significant differences relating to alloy content and treatment. Article 17.14 gives more information.

The tensile properties of metals are dependent upon the rate of straining, as shown for aluminum and copper in Fig. 17.4, and are significantly affected by the temperature as shown in Fig. 17.5. For high-temperature applications it is important to base

Table 17.1. Room-temperature Tensile Properties for Some Wrought Metals

Metal	Condition	Ultimate tensile σ_{ult}, psi	Yield strength σ_y, psi	% elongation	Modulus of elasticity, psi	Density, lb/cu in.
Aluminum (pure)..	Annealed	13,000	5,000	35	10,000,000	0.098
7075T6..........	Heat-treated	76,000	67,000	11	10,400,000	0.101
Copper (pure).....	Annealed	32,000	10,000	45	17,000,000	0.321
Cu-Be(2%)........	Heat-treated	200,000	150,000	2	19,000,000	0.297
Magnesium	Annealed	33,000	18,000	17	6,500,000	0.064
Mg-Al(8.5)A780A.	Strain relieved	44,000	31,000	18	6,500,000	0.065
Nickel (pure).....	Annealed	65,000	20,000	40	30,000,000	0.321
K Monel.........	Heat-treated	190,000	140,000	5	....	0.306
Titanium (pure)...	Annealed	59,000	40,000	28	15,000,000	0.163
Ti-Al(6)-V(14)α-β.	Heat-treated	170,000	150,000	7	15,000,000	0.163

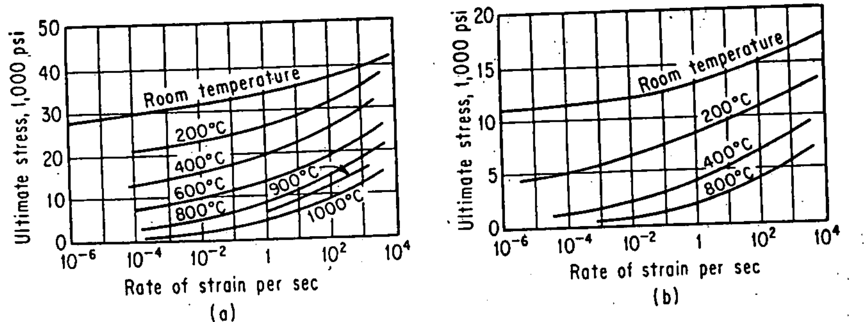


FIG. 17.4. Effects of strain rates and temperatures on tensile-strength properties of copper and aluminum.[1] (a) Copper. (b) Aluminum.

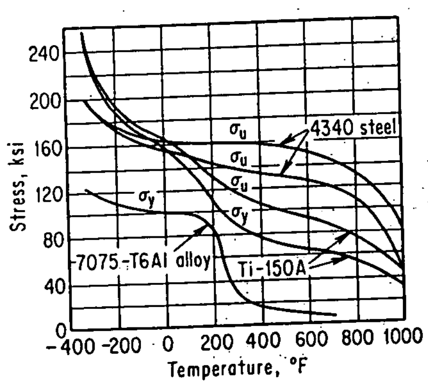


FIG. 17.5. Effects of temperatures on tensile properties.
σ_u = ultimate tensile strength
σ_y = yield strength

design on different criteria, notably the stress-rupture and creep characteristics in metals, both of which are also time-dependent phenomena. The use of metals at low temperatures requires a consideration of the possibility of brittleness, which can be measured in the impact test.

17.3. ATOMIC ARRANGEMENTS IN PURE METALS—CRYSTALLINITY

The basic structure of materials provides information upon which properties and behavior of metals may be generalized so that selection can be based on fundamental considerations. A regular and periodic array of atoms (in common metals whose atomic diameters are about one hundred-millionth of an inch) in space, in which a unit cell is the basic structure, is a fundamental characteristic of crystalline solids. Studies of these structures in metals lead to some important considerations of the behaviors

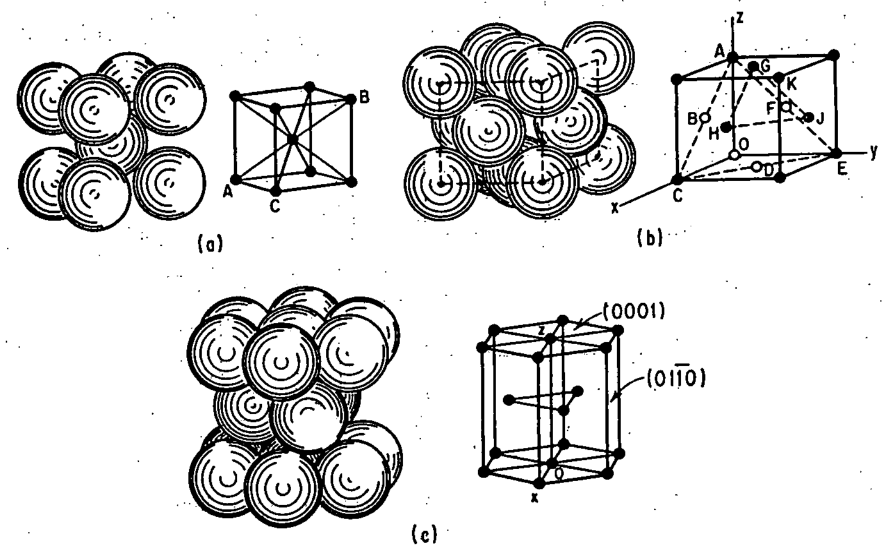


FIG. 17.6. Cell structure. (a) Body-centered cubic (b.c.c.) unit cell structure. (b) Face-centered cubic (f.c.c.) unit cell structure. (c) Hexagonal close-packed (h.c.p.) unit cell structure.

in response to externally applied forces, temperature changes, as well as applied electrical and magnetic fields.

The body-centered cubic (b.c.c.) cell shown in Fig. 17.6*a* is the atomic arrangement characteristic of αFe, W, Mo, Ta, βTi, V, and Nb. It is among this class of metals that transitions from ductile to brittle behavior as a function of temperature are significant to investigate. This structure represents an atomic packing density where about 66 per cent of the volume is populated by atoms while the remainder is free space. The elements Al, Cu, γFe, Ni, Pb, Ag, Au, and Pt have a closer packing of atoms in space constituting a face-centered cubic (f.c.c.) cell shown in Fig. 17.6*b*. Characteristic of these are ductility properties which in many cases extend to very low temperatures. Another structure, common to Mg, Cd, Zn, αTi, and Be, is the hexagonal close-packed (h.c.p.) cell in Fig. 17.6*c*. These metals are somewhat more difficult to deform plastically than the materials in the two other structures cited above.

It is apparent, from the atomic arrays represented in these structures, that the closest approach of atoms can vary markedly in different crystallographic directions. Properties in materials are anisotropic when they show significant variations in different

Table 17.2. Examples of Anisotropic Properties in Single Crystals

Property	Material and structure	Properties relation	Reference
Elastic modulus E in tension	αFe(b.c.c.)	$E_{(AB)} \sim 2.2E_{(AC)}$	Fig. 17.6a
Elastic modulus G in shear	Ag (f.c.c.)	$G_{(OC)} \sim 2.3G_{(OK)}$	Fig. 17.6b
Magnetization	αFe (b.c.c.)	Ease of magnetization	Fig. 17.7
Thermal expansion coefficient — α	Zn (h.c.p.)	$\alpha_{(OZ)} \sim 4\alpha_{(OA)}$	Fig. 17.6c

directions. Such tendencies are dependent on the particular structure and can be especially pronounced in single crystals (one orientation of the lattices in space). Some examples of these are given in Table 17.2. When materials are processed so that their final grain size is large (each grain represents one orientation of the lattices) or

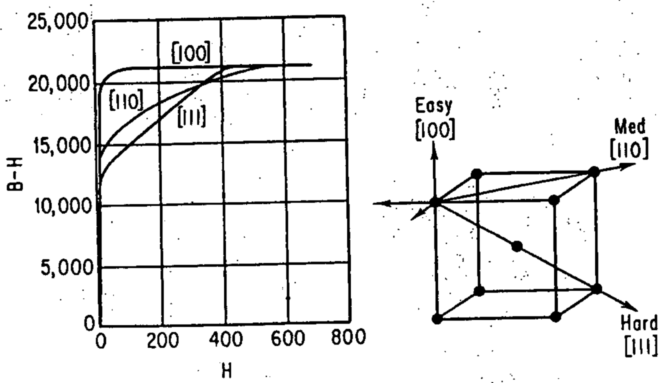


Fig. 17.7. Magnetic anisotropy in a single crystal of iron.[2]

$$I = \frac{B - H}{4\pi}$$

where I = intensity of magnetization
B = magnetic induction, gauss
H = field strength, oersteds

that the grains are preferentially oriented, as in extrusions, drawn wire, rolled sheet, sometimes in forgings and castings, special evaluation of anisotropy should be made. In the event that directional properties influence design considerations, particular attention must be given to metallurgical treatments which may control the degree of anisotropy. The magnetic anisotropy in a single crystal of iron is shown in Fig. 17.7.

17.4. PLASTIC DEFORMATION OF METALS

When metals are externally loaded past the elastic limit, so that permanent changes in shape occur, it is important to consider the induced internal stresses, property changes, and the mechanisms of plastic deformation. These are matters of practical consideration in the following: materials that are to be strengthened by cold work, machining of cold-worked metals, flow of metals in deep-drawing and impact extrusion operations, forgings where the grain flow patterns may affect the internal soundness, localized surface deformation to enhance fatigue properties, and cold working of some magnetic materials. Experimental studies provide the key by which important phenomena are revealed as a result of the plastic-deformation process. These studies

indicate some treatments that may be employed to minimize unfavorable internal-stress distributions and undesirable grain-orientation distributions.

Plastic deformation in metals occurs by a glide or slip process along densely packed planes fixed by the particular lattice structure in a metal. Therefore, an applied load is resolved as a shear stress, on those particular glide elements (planes and directions) requiring the least amount of deformation work on the system. An example of this

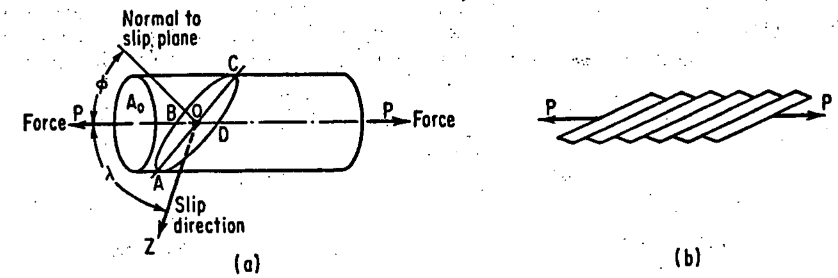


Fig. 17.8. Slip deformation in single crystals. (a) Resolved shear stress = $P/A_0 \cos\phi \cos\lambda$. $ABCD$ is plane of slip. OZ is slip direction. (b) Sketch of single crystal after yielding.

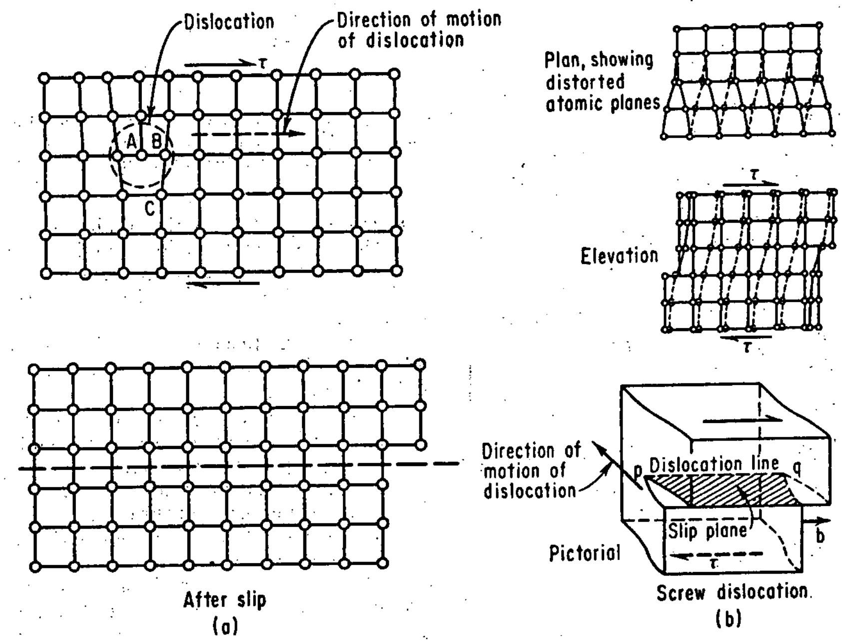


Fig. 17.9. Edge and screw dislocations as types of imperfections in metals.[2] (a) Edge dislocation. (b) Screw dislocation.

deformation process is shown in Fig. 17.8. Face-centered cubic structured metals, such as Cu, Al, and Ni, are more ductile than the hexagonal structured metals like Mg, Cd, and Zn at room temperature because in the f.c.c. structure there are four times as many possible slip systems as in a hexagonal structure. Slip is initiated at much lower stresses in metals than theoretical calculations based on a perfect array of atoms would indicate. In real crystals there are inherent structural imperfections termed dislocations (atomic misfits) as shown in Fig. 17.9, which account for the observed

yielding phenomenon in metals. In addition, dislocations are made mobile by mechanical and thermal excitations and they can interact to result in strain hardening of metals by cold work. Strength properties can be increased while the ductility is decreased in those metals which are amenable to plastic deformation. Cold working of pure metals and single-phase alloys provides the principal mechanism by which these may be hardened.

The yielding phenomenon is more nonhomogeneous in polycrystalline metals than in single crystals. Plastic deformation in polycrystalline metals initially occurs only in those grains in which the lattice axes are suitably oriented relative to the applied load axis, so that the critically resolved shear stress is exceeded. Other grains rotate and are dependent on the orientation relations of the slip systems and load application; these may deform by differing amounts.

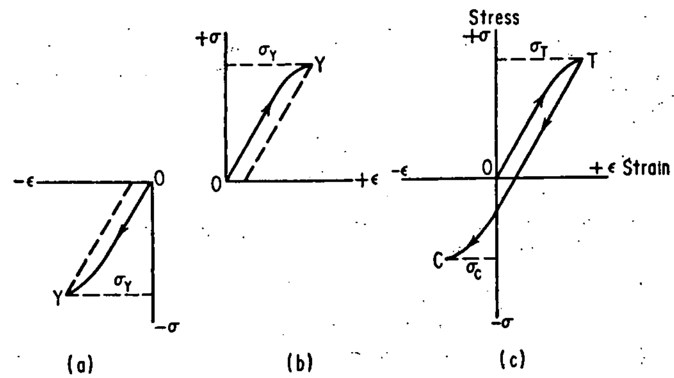


FIG. 17.10. The Bauschinger effect. (a) Compression. (b) Tension. The application of a compressive stress (a) or a tensile stress (b) results in the same value of yield strength σ_Y. (c) Stress reversal. A reversal of stress $O \rightarrow T \rightarrow C$ results in different values of tensile and compressive yield strengths; $\sigma_T \neq \sigma_C$.

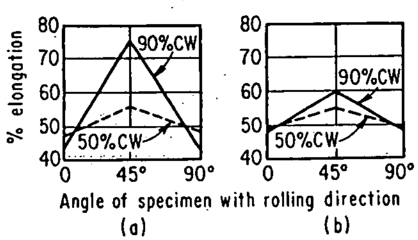


FIG. 17.11. Directionality in ductility in cold-worked and annealed copper sheet.[1] (a) Annealed at 1470°F. (b) Annealed at 750°F. The variation in ductility with direction for copper sheet is dependent on both the annealing temperature and the amount of cold work (per cent CW) prior to annealing.

As matters of practical considerations the following effects result from plastic deformation:

1. Materials become strain-hardened and the resistance to further strain hardening increases.

2. The tensile and yield strengths increase with increasing deformation, while the ductility properties decrease.

3. Macroscopic internal stresses are induced in which parts of the cross section are in tension while other regions have compressive elastic stresses.

4. Microscopic internal stresses are induced along slip bands and grain boundaries.

5. The grain orientations change with cold work so that some materials may exhibit different mechanical and physical properties in different directions.

The Bauschinger effect in metals is related to the differences in the tensile and compressive yield-strength values, as shown at σ_T and σ_C in Fig. 17.10 when a ductile metal undergoes stress reversal. This change in polycrystalline metals is the result of the nonuniform character of deformation and the different pattern of induced macrostresses. These grains, in which the induced macrostresses are compressive, will yield at lower values upon the application of a reversed compressive stress because they are already part way toward yielding. This effect is encountered in cold-rolled metals where there is lateral contraction together with longitudinal elongation; this accounts for the decreased yield strength in the lateral direction compared with the increased longitudinal yield strength.

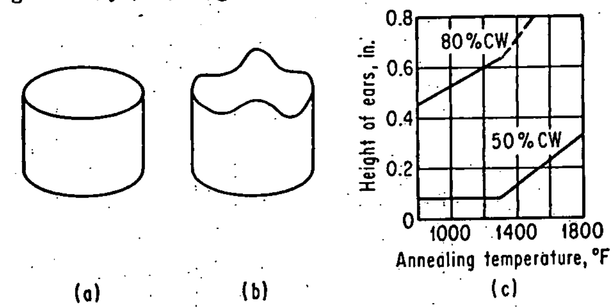


FIG. 17.12. The earing tendencies in cup deep-drawn from sheet. (a) Uniform flow, nonearing. (b) Eared cup, the result of nonuniform flow. (c) Height of ears in deep-drawn copper cups related to annealing temperature and amount of cold work.

The control of metal flow is important in deep-drawing operations performed on sheet metal. It is desirable to achieve a uniform flow in all directions. Cold-rolling sheet metal produces a structure in which the grains have a preferred orientation. This characteristic can persist, even though the metal is annealed (recrystallized), resulting in directional properties as shown in Fig. 17.11. A further consequence of this directionality, associated with the deep-drawing operation, is illustrated in Fig. 17.12. The important factors, involved with the control of earing tendencies, are the fabrication practices of the amount of cold work in rolling and duration and temperatures of annealing. When grain textural problems of this kind are encountered they can be studied by X-ray diffraction techniques and reasonably controlled by the use of optimum cold-working and annealing schedules.

17.5. PROPERTY CHANGES RESULTING FROM COLD-WORKING METALS

Cold-working metals by rolling, drawing, swaging, and extrusion is employed to strengthen them and/or to change their shape by plastic deformation. It is used principally on ductile metals which are pure, single-phase alloys and for other alloys which will not crack upon deformation. The increase in tensile strength accompanied by the decrease in ductility characteristic of this process is shown in Fig. 17.13. It is to be noted, especially from the yield-strength curve, that the largest rates of change occur during the initial amounts of cold reduction.

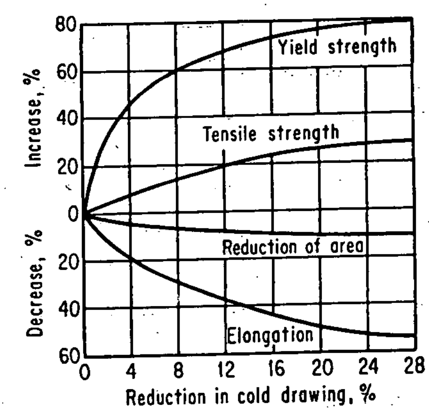


FIG. 17.13. Effect of cold drawing on the tensile properties of steel bars of up to 1 in. cross section having tensile strength of 110,000 psi or less before cold drawing.[3]

The variations in the macrostresses induced in a cold-drawn bar, illustrated in Fig. 17.14a, show that tensile stresses predominate at the surface. The equilibrium state of macrostresses throughout the cross section is altered by removing the surface layers in machining, the result of which may be warping in the machined part. It

may be possible, however, to stress-relieve cold-worked metals, which generally have better machinability than softened (annealed) metals, by heating below the recrystallization temperature. A typical alteration in the stress distribution, shown in Fig. 17.14*b*, is achieved so that the warping tendencies on machining are reduced, without decreasing the cold-worked strength properties. This stress-relieving treatment may also inhibit season cracking in cold-worked brasses subjected to corrosive environments

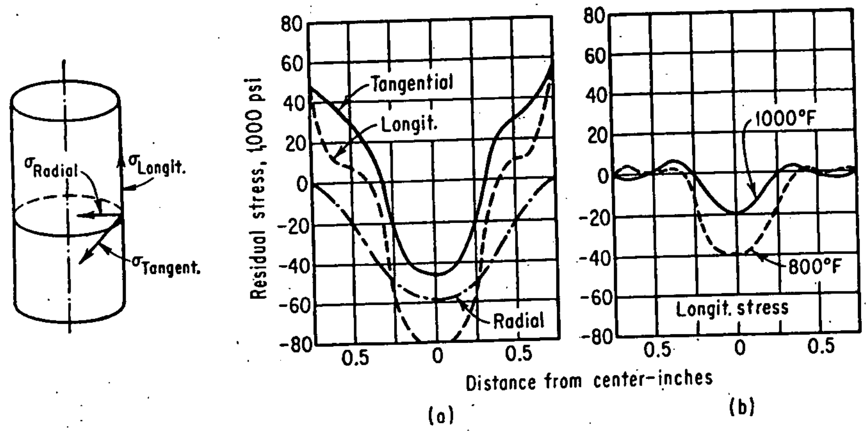


Fig. 17.14. Residual stress.[3] (*a*) In a cold-drawn steel bar 1½ in. in diameter 20 per cent cold drawn, 0.45C steel. (*b*) After stress-relieving bar.

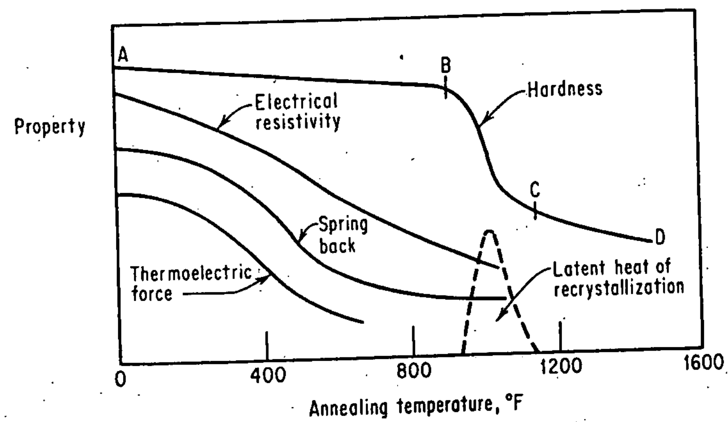


Fig. 17.15. The property changes in 95 per cent cold-worked iron with heating temperatures (1 hr). The temperature intervals:[4] $A \rightarrow B$, stress relieval, $B \rightarrow C$, recrystallization, and $C \rightarrow D$, grain growth, signify the important phenomena occurring.

containing amines. Since stressed regions, in a metal, are more anodic (i.e., go into solution more readily) than unstressed regions, it is often important to consider the relieving of stresses so that the designed member is not so likely to be subjected to localized corrosive attack.

Changes in electrical resistivity, elastic springback, and thermoelectric force, resulting from cold work, can be altered by a stress-relieval treatment, in a temperature range from *A* to *B*, as shown in Fig. 17.15. However, the grain flow pattern (preferred orientation) produced by cold working can be changed only by heating the metal to a temperature at which recrystallized stress-free grains will form.

Residual tensile stresses at the surface of a metal promote crack nucleation in the fatigue of metal parts. The use of a localized surface deformation treatment by shot peening, which induces compressive stresses in the surface fibers, offers the likelihood of improvement in fatigue and corrosion properties in alloys. Shot peening a forging flash line in high-strength aluminum alloys used in aircraft may also lessen the tendency toward stress-corrosion cracking. The effectiveness of this localized surface-hardening treatment is dependent on both the nature of surface discontinuities formed by shot peening and the magnitude of compressive stresses induced at the surface.

17.6. THE ANNEALING PROCESS

Metals are annealed in order to induce softening for further deformation, to relieve residual stresses, to alter the microstructure and, in some case after electroplating,

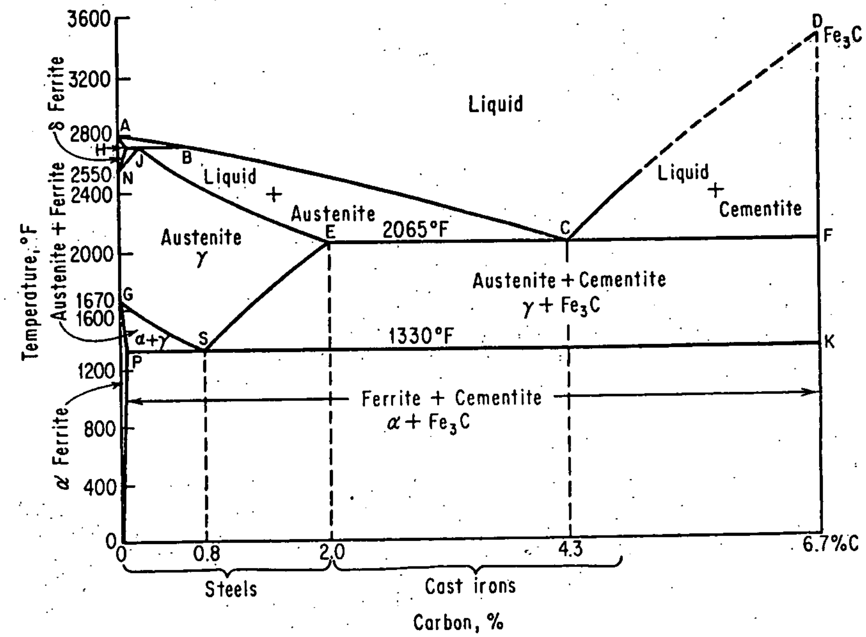


Fig. 17.16. The iron-carbon phase diagram.[4]

to expel, by diffusion, gases entrapped in the lattice. The process of annealing, the attaining of a strain-free recrystallized grain structure, is dependent mainly on the temperature, time, and the amount of prior cold work. The temperature indicated at *C* in Fig. 17.15 results in the complete annealing after 1 hr of the 95 per cent cold-worked iron. Heating beyond this temperature causes grains to grow by coalescence, so that the surface-to-volume ratio of the grains decreases together with decreasing the internal energy of the system. As the amount of cold work (from the originally annealed state) decreases, the recrystallization temperature increases and the recrystallized grain size increases. When a metal is cold-worked slightly (less than 10 per cent) and subsequently annealed, an undesirable roughened surface forms because of the abnormally large grain size (orange-peel effect) produced. These aspects of grain-size control in the annealing process enter in material specifications.

The annealing of iron-carbon-base alloys (steels) is accomplished by heating alloys of eutectoid and hypoeutectoid compositions (0.8 per cent C and less in plain carbon steels) to the single-phase region, austenite, as shown in Fig. 17.16 above the transition

line GS; and for hypereutectoid alloys (0.8 to 2.0 per cent C) between the transition lines SK and SE in Fig. 17.16; followed by a furnace cool at a rate of about 25°F per hour to below the eulectoid temperature SK. In the annealed condition, a desirable distribution of the equilibrium phases is thereby produced. A control of the micro-structure is manifested by this process in steels. Grain-size effects are principally controlled by the high-temperature treatment, grain sizes increasing with increasing temperatures, and in some cases minor impurity additions like vanadium inhibit grain coarsening to higher temperatures. These factors of grain-size control enter into the considerations of hardening steels by heat-treatment.

The control of the atmosphere in the annealing furnace is desirable in order to prevent gas-metal attack. Moisture-free neutral atmospheres are used for steels which oxidize readily. When copper and its alloys contain oxygen, as oxide, it is necessary to keep the hydrogen content in the atmosphere to a minimum. At temperatures lower than 900°F, the hydrogen should not exceed 1 per cent, and as the temperature is increased the hydrogen content should be reduced in order to prevent hydrogen embrittlement. In nickel and its alloys the atmosphere must be free from sulfur and slightly reducing by containing 2 per cent or more of CO. Some aluminum alloys containing magnesium are affected by high-temperature oxidation in annealing (and heat-treatment) and therefore require atmosphere control.

It is a characteristic property that strengths in all metals decrease with increasing temperatures. The coalescence of precipitate particles is one factor involved; so that material specifications for high-temperature use are concerned with alloy compositions that form particles having lower solubility and lower mobility. A second factor is concerned with the mobility of dislocations which increases at higher temperatures. Since strain hardening is reasoned to be due to the interaction of dislocations, then by the proper additions of solid-solution alloying elements that impede dislocations, resistance to softening will increase at the high temperature. The recrystallization temperature of iron is raised by the addition of 1 atomic per cent of Mn, Cr, V, W, Cb, Ta in the same order in which the atomic size of the alloying element differs from that of iron. The practical implications of these basic atomic considerations are important in selecting metals for high-temperature service.

17.7. THE PHASE DIAGRAM AS AN AID TO ALLOY SELECTION

Phase diagrams, which are determined experimentally and are based upon thermodynamic principles, are temperature-composition representations of slowly cooled alloys (annealed state). They are useful for predicting property changes with composition and selecting feasible fabrication processes. Phase diagrams also indicate the possible response of alloys to hardening by heat-treatment. Shown on these diagrams are first-order phase transitions and the phases present. In two-phase regions, the compositions of each phase are shown on the phase boundary lines and the relative amounts of each phase present can be determined by a simple lever relation at a given temperature.

The particular phases that are formed in a system are governed principally by the physical interactions of valency electrons in the atoms and secondarily by atomic-size factors. When two different atoms in the solid state exist on, or where one is in, an atomic lattice, the phase is a solid solution (e.g., γ austenite phase in Fig. 17.16) analogous to a miscible liquid solution. When the atoms are strongly electropositive and strongly electronegative to one another, an intermetallic compound is formed (e.g., Fe_3C, cementite). The two atoms are electronically indifferent to one another and a phase mixture issues (e.g., $\alpha + Fe_3C$) analogous to the immiscibility of water and oil.

The thermodynamic criteria for a first-order phase change, indicated by the solid lines on the phase diagram, are that, at the transition temperature, (1) the change in Gibbs's free energy for the system is zero; (2) there is a discontinuity in entropy (a latent heat of transformation and a discontinuous change in specific heat); and (3) there is a discontinuous change in volume (a dilational effect).

In the selection of alloys for sound castings, particular attention is given that part

of the system where the liquidus line ($ABCD$) goes through a minimum. For alloys between the composition limits of $E \rightarrow F$, a eutectic reaction occurs at 2065°F such that

$$\text{liq } C = \gamma + Fe_3C$$

It is for this reason in the iron-carbon system that cast irons are classified as having carbon contents greater than 2 per cent. For purposes of controlling grain size, obtaining sound castings free from internal porosities (blowholes) and internal shrinkage cavities, and possessing good mold-filling characteristics, alloys and low-melting solutions are chosen near the eutectic composition (i.e., at C). Aluminum-silicon die-casting alloys have a composition of about 11 per cent silicon near the eutectic composition. Special considerations need be given to the properties and structures in cast irons because the Fe_3C phase is thermodynamically unstable and decomposition to graphite (in gray cast irons) may result.

The predominant phase-diagram characteristic in steels is the eutectoid reaction, in the solid state, along GSE where $\gamma_s = \alpha + Fe_3C$ (pearlite) at 1330°F. Steels are therefore classified as alloys in the Fe-C system having a carbon content less than 2.0 per cent C; and furthermore, according to their applications, compositions are designated as hypoeutectoid ($C < 0.8$ per cent), eutectoid ($C = 0.8$ per cent), and hypereutectoid ($C < 2$ per cent > 0.8 per cent). Since the slowly cooled room-temperature structures of steels contain a mechanical aggregate of the ferrite and Fe_3C-cementite phases, the property relations vary linearly as shown in Fig. 17.17. The ductility decreases with increasing carbon contents.

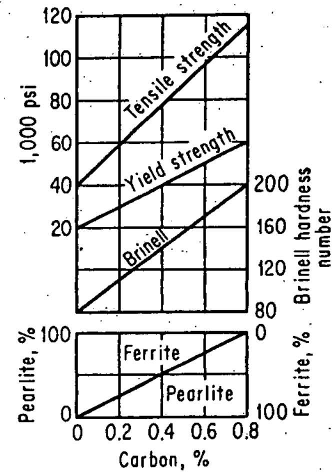


FIG. 17.17. Relation of mechanical properties and structure to carbon content of slowly cooled carbon steels.[4]

Some important characteristics of the equilibrium phases in steels are:

Phase	*Characteristics*
α ferrite	Low C solubility (less than 0.03%) body-centered cubic, ductile and ferromagnetic below 1440°F
Fe_3C, cementite	Intermetallic compound, orthorhombic, hard, brittle, and fixed composition at 6.7% C
γ austenite	Can dissolve up to 2% C in solid solution, face-centered cubic, nonmagnetic, and in this region annealing, hardening, forging, normalizing, and carburizing processes take place

Low-carbon alloys can be readily worked by rolling, drawing, and stamping because of the predominant ductility of the ferrite. Wires for suspension cables having a carbon content of about 0.7 per cent are drawn at about 1100°F (patenting) because of the greater difficulty, in room-temperature deformation, caused by the presence of a relatively large amount of the brittle Fe_3C phase.

Extensive substitutional solid-solution alloys form in binary systems when they have similar chemical characteristics and atomic diameters in addition to having the same lattice structure. Such alloys include copper-nickel (monel metal being a commercially useful one), chromium-molybdenum, copper-gold, and silver-gold (jewelry alloys). The phase diagram and the equilibrium-property changes for this system are shown in Fig. 17.18*a*. Each pure element is strengthened by the addition of the other whereby the strongest alloy is at an equal atom concentration. There are no first-order phase changes up to the start of melting (the solidus line EHG); so that these are not hardened by heat-treatment but only by cold work. The electrical conductivity decreases from each end of the composition axis. Because of the presence of but one

phase, these alloys are selected for their resistance to electrochemical corrosion. High-temperature-service metals are alloys which have essentially a single-phase solid solution with minor additions of other elements to achieve specific effects.

Another important system is one in which there are present regions of partial solid solubility as shown in Fig. 17.18*b* together with equilibrium-property changes. An important consideration in the selection of alloys containing two or more phases is that galvanic-corrosion attack may occur when there exists a difference in the electromotive potential between the phases in the environmental electrolyte. Sacrificial galvanic protection of the base metal in which the coating is more anodic than the base metal is used in zinc-plating iron-base alloys (galvanizing alloys). The intimate mechanical mixture of phases which are electrochemically different may result in pitting corrosion, or even more seriously, intergranular corrosion may result if the alloy is improperly treated by causing localized precipitation at grain boundaries.

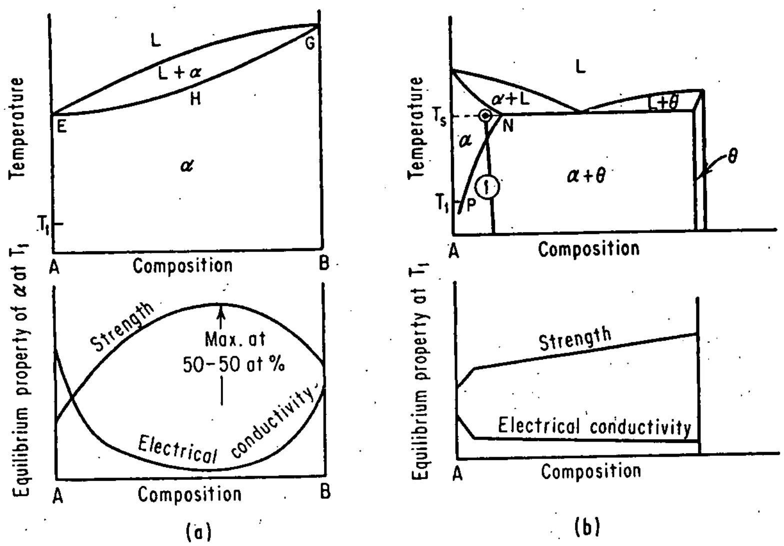


FIG. 17.18. Binary systems.[4] (*a*) Complete solid-solubility phase diagram. (*b*) Partial solid-solubility part of phase diagram. α is a substitutional solid solution, a phase with two different atoms on the same lattice. In the AlCu system θ is an intermediate phase (precipitant) having a composition nominally of $CuAl_2$.

Heat-treatment by a precipitation-hardening process is indeed an important strengthening mechanism in particular alloys such as the aircraft aluminum-base, copper-beryllium, magnesium-aluminum, and alpha-beta titanium alloys (Ti, Al, and V). In these alloys a distinctive feature is that the solvus line *NP* in Fig. 17.18*b* shows decreasing solid solubility with decreasing temperature. This in general is a necessary, but not necessarily sufficient, condition for hardening by precipitation since other thermodynamic conditions as well as coherency relations between the precipitated phases must prevail. The sequence of steps for this process is as follows: An alloy is solution heat-treated to a temperature T_s, rapidly quenched so that a metastable supersaturated solid solution is attained, and then aged at experimentally determined temperature-time aging treatments to achieve desired mechanical properties. This is the principal hardening process for those particular nonferrous alloys (including Inconels) which can respond to a precipitation-hardening process.

The engineer is frequently concerned with the strength-to-density ratio of materials and its variation with temperature. A number of materials are compared on this

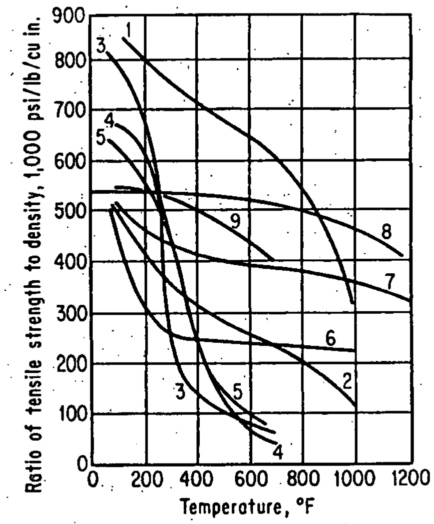


FIG. 17.19. Approximate comparison of materials on a strength-weight basis from room temperature to 1000°F.[6]

1. T_1 – 8Mn
2. 9990 T_1
3. 75S – T6Al
4. 24S – T4Al
5. AZ31A Mg
6. Annealed stainless steel
7. Half-hard stainless steel
8. Inconel X
9. Glass-cloth laminate

basis in Fig. 17.19 in which the alloys designated by curves 1, 3, 4, 5, and 8 are heat-treatable nonferrous alloys.

17.8. HEAT-TREATMENT CONSIDERATIONS FOR STEEL PARTS

The heat-treating process for steel involves heating to the austenite region where the carbon is soluble, cooling at specific rates, and tempering to relieve some of the stress which results from the transformation. Some important considerations involved in specifying heat-treated parts are strength properties, warping tendencies, mass effects (hardenability), fatigue and impact properties, induced transformation stresses, and the use of surface-hardening processes for enhanced wear resistance. Temperature and time factors affect the structures issuing from the decomposition of austenite; for a eutectoid steel (0.8 per cent C) they are:

Decomposition product from γ	Structure	Mechanism	Temp. range, °F
Pearlite........	Equilibrium ferrite + Fe_3C	Nucleation; growth	1300–1000
Bainite........	Nonequilibrium ferrite + carbide	Nucleation; growth	1000–450
Martensite.....	Supersaturated tetragonal lattice	Diffusionless	M_s(450 + lower)

The tensile strength of a slowly cooled (annealed) eutectoid steel containing a coarse pearlite structure is about 120,000 psi. To form bainite, the steel must be cooled rapidly enough to escape pearlite transformation and must be kept at an intermediate temperature range to completion, from which a product having a tensile

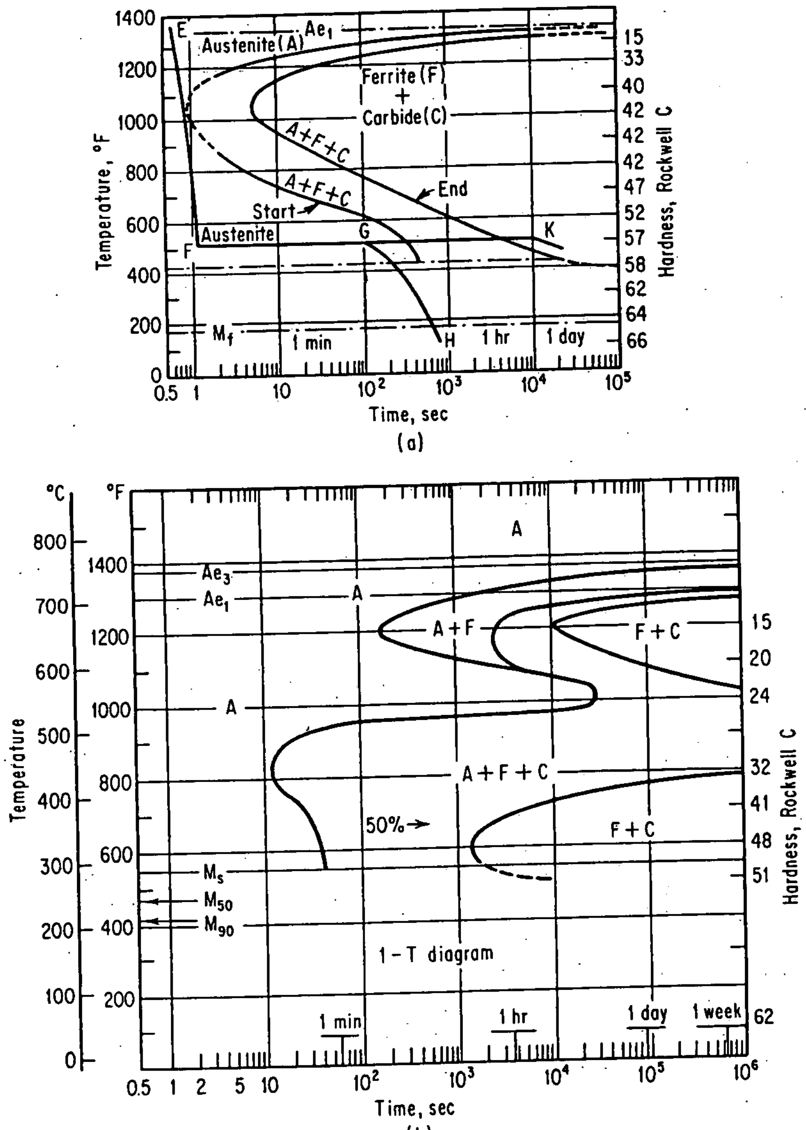


Fig. 17.20. Isothermal transformation diagram.[5] (a) Eutectoid carbon steel. (b) SAE 4340 steel.

M_S = start of martensite temperature
M_f = finish of martensite temperature
$EFGH$ = martempering (follow by tempering)
EFK = austempering

A—Austenite; F—Ferrite; C—Carbide
M_s and M_f—temperatures for start and finish of Martensile transformation
M_{50} and M_{90}—temperatures for 50% and 90% Martensile transformation

strength of about 250,000 psi can be formed. Martensite, the hardest and most brittle product, forms independently of time by quenching rapidly enough to escape higher-temperature transformation products. The carbon atoms are trapped in the martensite, causing its lattice to be highly strained internally; its tensile strength is in excess of 300,000 psi. Isothermal transformation characteristics of all steels show the temperature-time and transformation products as in Fig. 17.20, where the lines indicate the start and end of transformation. On the temperature-time coordinates involved cooling curves can be superimposed, which show that, for a 1-in. round water-quenched specimen, mixed products will be present. The outside will be martensite and the middle sections will contain pearlite. Alloying elements are added to steels principally to retard pearlite transformation either so that less drastic quenching media can be used or to ensure more uniform hardness throughout. This retardation is shown in Fig. 17.20b for a SAE 4340 steel containing alloying additions of Ni, Cr, or Mo and 0.4 per cent C.

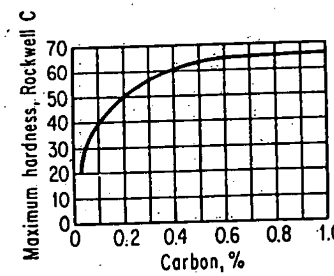


Fig. 17.21. Relation of maximum attainable hardness of quenched steels to carbon content.[7]

The carbon content in steels is the most significant element upon which selection for the maximum attainable hardness of the martensite is based. This relation is shown in Fig. 17.21. Since the atomic rearrangements involved in the transformation from the face-centered-cubic austenite to the body-centered-tetragonal martensite result in a volumetric expansion, on cooling, of about 1 per cent (for a eutectoid steel) nonuniform stress patterns can be induced on transformation. As cooling starts at the surface, by the normal process of heat transfer, parts of a member can be expanding, because of transformation, while further inward normal contraction occurs on the cooling austenite. The danger of cracking and distortion (warping) as a consequence of the steep thermal gradients and the transformation involved in hardening steels can be eliminated by using good design and the selection of the proper alloys. Where section size, time factors, and alloy content (as it affects transformation curves) permit, improved practices by martempering shown by $EFGH$ in Fig. 17.20, followed by tempering or austempering shown by EFK, may be feasible and are worthy of investigation for the particular alloy used.

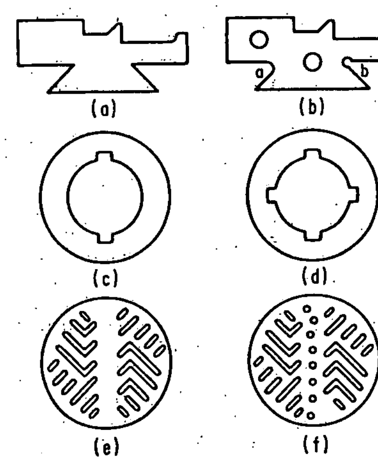


Fig. 17.22. Examples of good (b, d, and f) and bad (a, c, and e) designs for heat-treated parts.[8] (1) b is better than a because of fillets and more uniform mass distribution. (2) In c, cracks may form at keyways. (3) Warping may be more pronounced in e than in f, which are blanking dies.[8]

Uniform mass distribution, and the elimination of sharp corners (potential stress raisers) by the use of generous fillets, are recommended. Some design features pertinent to the elimination of quench cracks and the minimization of distortion by warping are illustrated by Fig. 17.22 in which a, c, and e represent poor designs in comparison with the suggested improvements apparent in b, d, and f.

Steels are tempered to relieve stresses, to impart ductility, and to produce a desirable microstructure by a reheating process of the quenched member. The tempering process is dependent on the temperature, time, and alloy content of the steel. Differ-

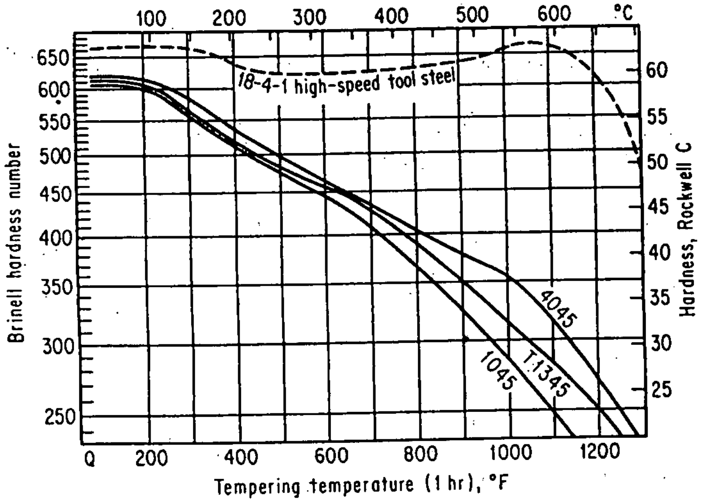


FIG. 17.23. Effect of tempering temperature on the hardnesses of SAE 1045, T1345, and 4045 steels. In the high-speed tool steel 18-4-1 secondary hardening occurs at about 1050°F.[9]

ent alloys soften at different rates according to the constitutionally dependent diffusional structure. The response to tempering for 1 hr for three different steels of the same carbon content is shown in Fig. 17.23. In addition, the tempering characteristics of a high-speed tool steel, 18 per cent W, 4 per cent Cr, 1 per cent V, and 0.9 per cent C, are shown to illustrate the secondary hardening at about 1050°F. The pronounced tendency for high-carbon steels to retain austenite on transformation normally has deleterious effects on dimensional stability and fatigue performance. In high-speed tool steel, the secondary hardening is due to the transformation of part of the retained austenite to newly transformed martensite. The structure contains tempered and untempered martensite with perhaps some retained austenite. Multiple tempering treatments on this type of steel produce a more uniform product.

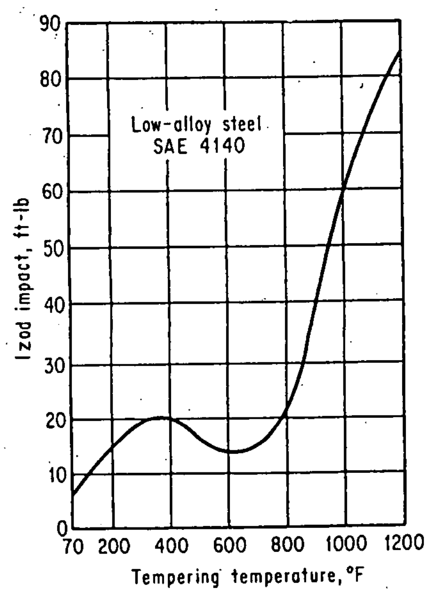


FIG. 17.24. In the tempering of this 4140 steel the notched-bar impact properties decrease in the range of 450 to 650°F.[9]

In low-alloy steels where the carbon content is above 0.25 per cent, there may be a tempering-temperature interval at about 450 to 650°F, during which the notch impact strength goes through a minimum. This is shown in Fig. 17.24 and is associated with the formation of an embrittling carbide network (ϵ carbide) about the martensite subgrain boundaries. Tempering is therefore carried out up to 400°F where the parts are to be used principally for wear resistance, or in the range of 800 to 1100°F where greater toughness is required. In the nomenclature of structural steels, adopted by the Society of Automotive Engineers and the American Iron

and Steel Institute, the first two numbers designate the type of steel according to the principal alloying elements and the last two numbers designate the carbon content:

Series designation	*Types*
10xx	Nonsulfurized carbon steels
11xx	Resulfurized carbon steels (free-machining)
12xx	Rephosphorized and resulfurized carbon steels (free-machining)
13xx	Manganese 1.75%
23xx*	Nickel 3.50%
25xx*	Nickel 5.00%
31xx	Nickel 1.25%, chromium 0.65%
33xx	Nickel 3.50%, chromium 1.55%
40xx	Molybdenum 0.20 or 0.25%
41xx	Chromium 0.50 or 0.95%, molybdenum 0.12 or 0.20%
43xx	Nickel 1.80%, chromium 0.50 or 0.80%, molybdenum 0.25%
44xx	Molybdenum 0.40%
45xx	Molybdenum 0.52%
46xx	Nickel 1.80%, molybdenum 0.25%
47xx	Nickel 1.05%, chromium 0.45%, molybdenum 0.20 or 0.35%
48xx	Nickel 3.50%, molybdenum 0.25%
50xx	Chromium 0.25, 0.40, or 0.50%
50xxx	Carbon 1.00%, chromium 0.50%
51xx	Chromium 0.80, 0.90, 0.95, or 1.00%
51xxx	Carbon 1.00%, chromium 1.05%
52xxx	Carbon 1.00%, chromium 1.45%
61xx	Chromium 0.60, 0.80, or 0.95%, vanadium 0.12%, 0.10% min, or 0.15% min
81xx	Nickel 0.30%, chromium 0.40%, molybdenum 0.12%
86xx	Nickel 0.55%, chromium 0.50%, molybdenum 0.20%
87xx	Nickel 0.55%, chromium 0.50%, molybdenum 0.25%
88xx	Nickel 0.55%, chromium 0.50%, molybdenum 0.35%
92xx	Manganese 0.85%, silicon 2.00%, chromium 0 or 0.35%
93xx	Nickel 3.25%, chromium 1.20%, molybdenum 0.12%
94xx	Nickel 0.45%, chromium 0.40%, molybdenum 0.12%
98xx	Nickel 1.00%, chromium 0.80%, molybdenum 0.25%

* Not included in the current list of standard steels.

The most probable properties of tempered martensite for low-alloy steels fall within narrow bands even though there are differences in sources and treatments. The relations for these shown in Fig. 17.25 are useful in predicting properties to within approximately 10 per cent.

Structural steels may be specified by hardenability requirements, the H designation, rather than stringent specification of the chemistry. Hardenability, determined by the standardized Jominy end-quench test, is a measurement related to the variation in hardness with mass, in quenched steels. Since different structures are formed as a function of the cooling rate and the transformation is affected by the nature of the alloying elements, it is necessary to know whether the particular steel is shallow (A) or deeply hardenable (C), as in Fig. 17.26. The hardenability of a particular steel is a useful criterion in selection because it is related to the mechanical properties pertinent to the section size.

The selection of through-hardened steel based upon carbon content is indicated below for some typical applications.

Carbon range	Requirement	Approx tensile strength level, psi	Applications
Medium, 0.3 to 0.5%	Strength and toughness	150,000	Shafts, bolts, forgings, nuts
Intermediate, 0.5 to 0.7%	Strength	225,000	Springs
High, 0.8 to 1.0%	Wear resistance	Greater than 300,000	Bearings, rollers, bushings

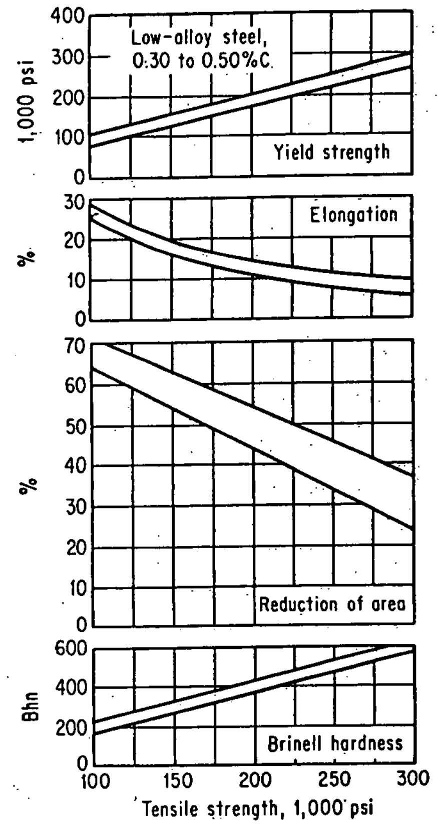


FIG. 17.25. The most probable properties of tempered martensite for a variety of low-alloy steels.[4]

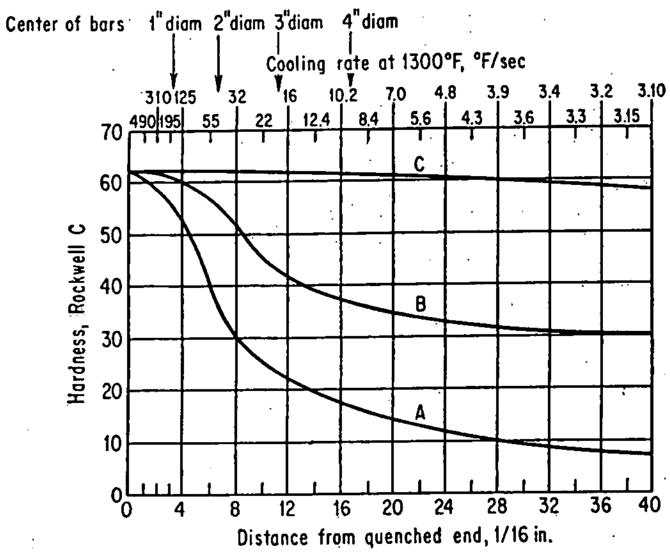


FIG. 17.26. Hardenability curves for different steels with the same carbon content.[7]

17.9. SURFACE-HARDENING TREATMENTS

The combination of high surface wear resistance and a tough-ductile core is particularly desirable in gears, shafts, and bearings. Various types of surface-hardening treatments and processes can achieve these characteristics in steels; the most important of these are the following:

Base metal	*Process*
Low C to 0.3%	Carburization—A carbon diffusion in the γ-phase region, with controlled hydrocarbon atmosphere or in a box filled with carbon. The case depth is dependent on temperature-time factors. Heat-treatment follows process
Medium C, 0.4 to 0.5%	Localized surface heating by induction or a controlled flame to above the Ac_3 temperature; quenched and tempered
Nitriding (Nitralloys, stainless steels)	Formation of nitrides (in heat-treated parts) in ammonia atmosphere at 950 to 1000°F, held for long times. A thin and very hard surface forms and there may be dimensional changes
Low C, 0.2%	Cyaniding—parts placed in molten salt baths at heat-treating temperatures; some limited carburization and nitriding occur for cases not exceeding 0.020 in. Parts are quenched and tempered

The carbon penetration in carburization is determined by temperature-time-distance relations issuing from the solution of diffusional equations where D, the diffusion coefficient, is independent of concentrations. These relations, shown in Fig. 17.27, permit the selection of a treatment to provide specific case depths. Typical applications are as follows:

Case depth, in.	*Applications (automotive)*
More than 0.020	Push rods, light-load gears, water pump shafts
0.020–0.040	Valve rocker arms, steering-arm bushings, brake and clutch pedal shafts
0.040–0.060	Ring gears, transmission gears, piston pins, roller bearings
Greater than 0.060	Camshafts

The heat-treatments used on carburized parts depend upon grain-size requirements, minimization of retained austenite in the microstructure, amount of undissolved carbide network, and core-strength requirements. As a result of carburization, the surface fiber stresses are compressive. This leads to better fatigue properties. This treatment, which alters the surface chemistry by diffusion of up to 1 per cent carbon in a low-carbon steel, gives better wear resistance because the surface hardness is treated for values above 60 Rockwell C, while the low-carbon-content core has ductile properties to be capable of the transmission of torsional or bending loads.

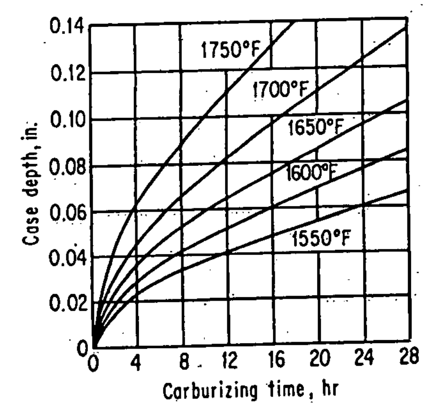


FIG. 17.27. Relation of time and temperature to carbon penetration in gas carburizing.[10]

Selection of the nitriding process requires careful consideration of cost because of the long times involved in the case formation. A very hard case having a hardness of about 70 Rockwell C ensures excellent wear resistance. Nitrided parts have good corrosion resistance and improved fatigue properties. Nitriding follows the finish

machining and grinding operations, and many parts can be nitrided without great likelihood of distortion. Long service (several hundred hours) at 500°F has been attained in nitrided gears made from a chromium-base hot-work steel H11. Some typical nitrided steels and their applications are as follows:

Steel	Nitriding treatment		Case hardness, in.	Case depth Rockwell C	Applications
	Hr	°F			
4140	48	975	0.025–0.035	53–58	Gears, shafts, splines
4340	48	975	0.025–0.035	50–55	Gears, drive shafts
Nitralloy 135M	48	975	0.020–0.025	65–70	Valve stems, seals, dynamic faceplates
H11	70	960 + 980	0.015–0.020	67–72	High-temperature power gears, shafts, pistons

17.10. NOTCHED IMPACT PROPERTIES—CRITERIA FOR MATERIAL SELECTION

When materials are subject to high deformation rates and are particularly sensitive to stress concentrations at sharp notches, criteria must be established to indicate safe operating-temperature ranges. The impact test (Izod or Charpy V-notch) performed on notched specimens conducted over a prescribed temperature range indicates the likelihood of ductile (shear-type) or brittle (cleavage-type) failure. In this test the velocity of the striking head at the instant of impact is about 18 fps; so that the strain rates are several orders of magnitude greater than in a tensile test. The energy absorbed in fracturing a standard notched specimen is measured by the differences in potential energy from free fall of the hammer to the elevation after fracturing it. The typical effect of temperature upon impact energy for a metal which shows ductile and brittle characteristics is shown in Fig. 17.28. Interest centers on the transition temperature range and the material-sensitive factors such as composition, microstructure, and embrittling treatments. ASTM specifications for structural steel for ship plates specify a minimum impact energy at a given temperature, as, for example, 15 ft-lb at 40°F. It is desirable to use materials at impact energy levels and at service temperatures where crack propagation does not proceed. Some impact characteristics for construction steels are shown in Fig. 17.29.

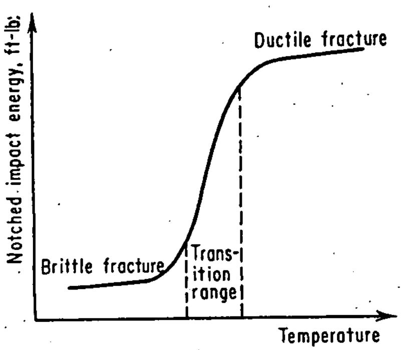


FIG. 17.28. The ductile-to-brittle transition in impact.

It is generally characteristic of pure metals which are face-centered cubic in lattice structure to possess toughness (have no brittle fracture tendencies) at very low temperatures. Body-centered-cubic pure metals, as well as hexagonal metals, do show ductile-to-brittle behavior. Tantalum, a body-centered-cubic metal, is a possible exception and is ductile in impact even at cryogenic temperatures. Alloyed metals do not follow any general pattern of behavior; some specific impact values for these are as follows:

Material	Charp V-notch at 80°F	Impact strength, ft-lb at −100°F
Cu-Be(2%) HT	5.4	5.5
Phosphor bronze (5% Sn):		
Annealed	167	193
Spring temper	46	44
Nilvar Fe-Ni (36%):		
Annealed	218	162
Hard	97	77
2024-T6 aluminum aircraft alloy, HT aged	12	12
7079-T6 aluminum aircraft alloy, HT aged	4.5	3.5
Mg-Al-Zn extruded	7.0	3.0

Austenitic stainless steels are ductile and do not exhibit transition in impact down to very low temperatures.

Some brittle service failures in steel structures have occurred in welded ships, gas-transmission pipes, pressure vessels, bridges, turbine generator rotors, and storage tanks. Serious consideration must then be given the effect of stress raisers, service temperatures, tempering embrittling structures in steels, grain-size effects, as well as the effects of minor impurity elements in materials.

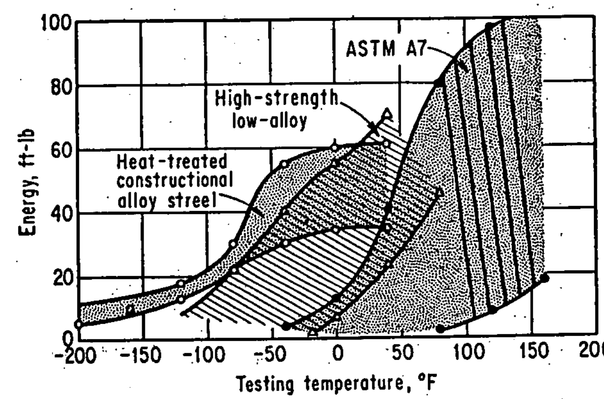


FIG. 17.29. Influence of testing temperature on notch toughness, comparing carbon steel of structural quality (ASTM A7) with high-strength low-alloy and heat-treated constructional alloy steel.[4] Charpy V-notch.

17.11. FATIGUE CHARACTERISTICS FOR MATERIALS SPECIFICATIONS

Most fatigue failures observed in service as well as under controlled laboratory tests are principally the result of poor design and machining practice. The introduction of potential stress raisers by inadequate fillets, sharp undercuts, and toolmarks at the surfaces of critically cyclically stressed parts may give rise to crack nucleation and propagation so that ultimate failure occurs. Particular attention should be given to material fatigue properties where rotating and vibrating members experience surface fibers under reversals of stress.

In a fatigue test, a highly polished round standard specimen is subjected to cyclic

loading; the number of stress reversals to failure is recorded. For sheets, the standard specimen is cantilever supported. Failure due to tensile stresses usually starts at the surface. Typical of a fatigue fracture is its conchoidal appearance, where there is a smooth region in which the severed sections rubbed against each other and where the crack progressed to a depth where the load could no longer be sustained. From the fatigue data a curve of stress vs. number of cycles to failure is plotted. Note that there can be considerable statistical fluctuation in the results (about ±15 per cent variations in stress).

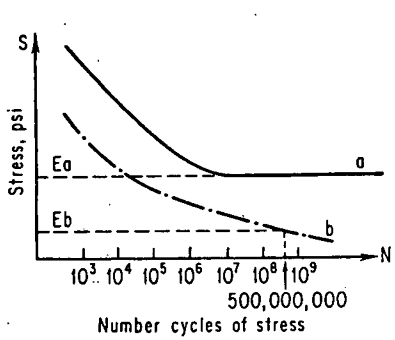


FIG. 17.30. Fatigue curves. (*a*) Most materials have an endurance limit E_a (asymptotic stress). (*b*) Endurance limit E_b (non-asymptotic) set at arbitrary value of *N*.

Two characteristics can be observed in fatigue curves with respect to the endurance limit shown by E_a and E_b in Fig. 17.30:

E_a, curve *A*, the asymptotic stress value, typical in most materials

E_b, curve *B*, a stress value taken at an arbitrary number of cycles; e.g., 500,000,000, typical in Al and Mg alloys

For design specifications, the endurance limit represents a safe working stress for fatigue. The endurance ratio is defined as the ratio of the endurance limit to the ultimate tensile strength. These values are strongly dependent upon the presence of notches on the surface and a corrosive environment, and on surface-hardening treatments. In corrosion, the pits formed act as stress raisers leading generally to greatly reduced endurance ratios. References 15 through 20 provide useful information on corrosion. A poorly machined surface or a rolled sheet with surface scratches evidences low endurance ratios, as do parts in service with sharp undercuts and insufficiently filleted changes in section.

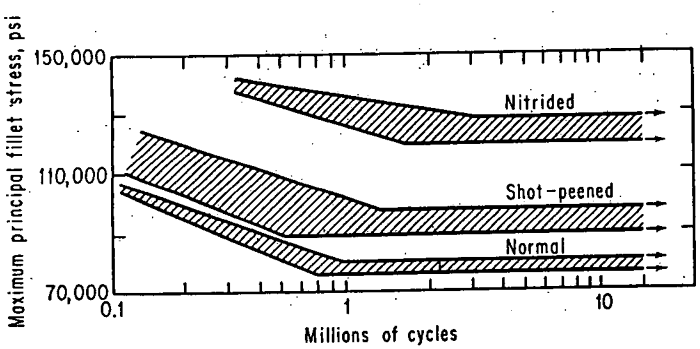


FIG. 17.31. Effects of surface-hardening treatments in improving the fatigue characteristics of a 4340 steel used as an aircraft crankshaft.[11]

Improvements in fatigue properties are brought about by those surface-hardening treatments which produce induced compressive stresses as in steels, nitrided (about 160,000 psi compressive stress) or carburized (about 35,000 psi compressive stress). An example of the effects of surface-hardening treatments on an aircraft crankshaft made from a 4340 hardened and tempered steel (30 Rockwell C) is shown in Fig. 17.31.

Metallurgical factors related to poorer fatigue properties are the presence of retained austenite in hardened steels, the presence of flakes or sharp inclusions in the

microstructure, and treatments which induce preferential corrosive grain-boundary attack. When parts are quenched or formed so that surface tensile stresses are present, stress-relieval treatments are advisable.

17.12. MATERIALS FOR HIGH-TEMPERATURE APPLICATIONS

(a) Introduction

Selection of materials to withstand stress at high temperature is based upon experimentally determined temperature stress-time properties. Some useful engineering design criteria follow.

1. Dimensional change, occurring by plastic flow, when metals are stressed at high temperatures for prolonged periods of time, as measured by creep tests
2. Stresses that lead to fracture, after certain set time periods, as determined by stress-rupture tests, where the stresses and deformation rates are higher than in a creep test
3. The effect of environmental exposure on the oxidation or scaling tendencies
4. Considerations of such properties as density, melting point, emissivity, ability to be coated and laminated, elastic modulus, and the temperature dependence on thermal conductivity and thermal expansion

Furthermore, the microstructural changes occurring in alloys used at high temperatures are correlated with property changes in order to account for the significant discontinuities which occur with exposure time. As a result of these evaluations, special alloys that have been (or are being) developed are recommended for use in different temperature ranges extending to about 2800°F (refractory range). Vacuum or electron-beam melting and special welding techniques are of special interest here in fabricating parts.

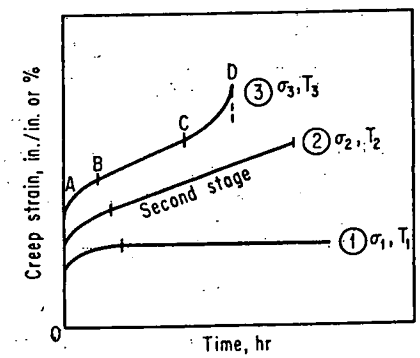


FIG. 17.32. Typical creep curves. At constant temperature, $\sigma_3 > \sigma_2 > \sigma_1$. At constant stress, $T_3 > T_2 > T_1$.

(b) Creep and Stress—Rupture Properties

In a creep test, the specimen is heated in a temperature-controlled furnace, an axial load is applied, and the deformation is recorded as a function of time, for periods of 1,000 to 3,000 hr. Typical changes in creep strain with time, for different conditions of stress and temperature, are shown in Fig. 17.32. Plastic flow creep, associated with the movement of dislocations by climb sliding of grain boundaries and the diffusion of vacancies, is characterized by:

1. *OA*, elastic extension on application of load
2. *AB*, first stage of creep with changing rate of creep strain
3. *BC*, second stage of creep, in which strain rate is linear and essentially constant
4. *CD*, third stage of increasing creep rate leading to fracture

Increasing stress at a constant temperature or increasing temperature at constant stress results in the transfers from the 1 to 2 to 3 curves in Fig. 17.32.

The engineering design considerations for dimensional stability are based upon

1. Stresses resulting in a second-stage creep rate of 0.0001 per cent per hour (1 per cent per 10,000 hr or 1 per cent per 1.1 years)
2. A second-stage creep rate of 0.00001 per cent per hour (1 per cent per 100,000 hr

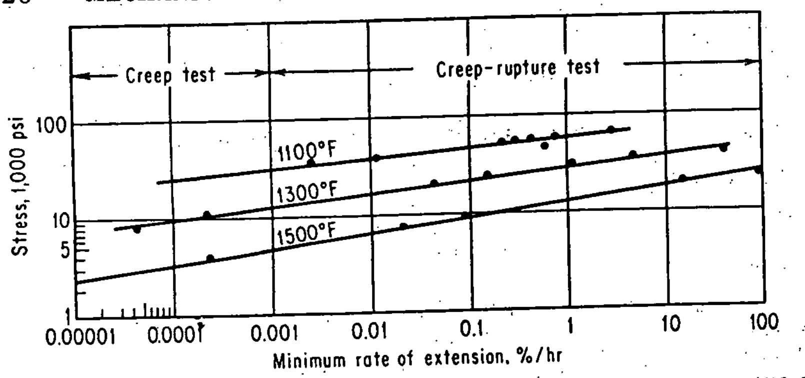


FIG. 17.33. Correlation of creep and rupture test data for type 316 stainless steel (18 Cr, 8 Ni, and Mo).[12]

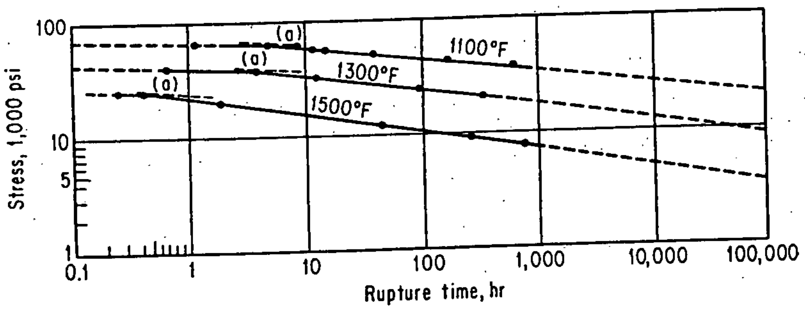


FIG. 17.34. Stress vs. rupture time for type 316 stainless steel.[12] The structural character associated with point (*a*), on each of the three relations, is that the mode of fracture changes from transgranular to intergranular.

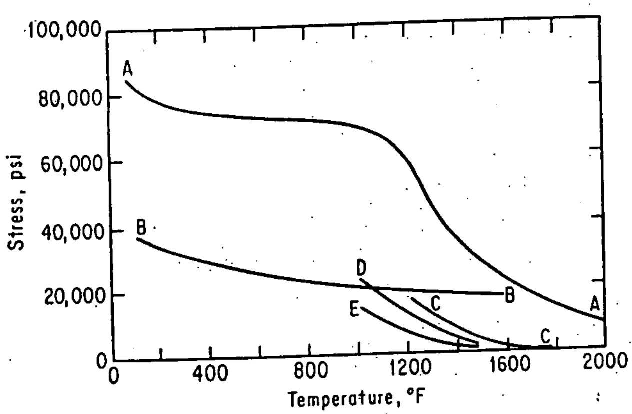


FIG. 17.35. Properties of type 316 stainless steel (18 Cr, 8 Ni, and Mo).[12]

A = short-time tensile strength
B = short-time yield strength, 0.2 per cent offset
C = stress for rupture 10,000 hr
D = stress for creep rate 0.0001 per cent per hr
E = stress for creep rate 0.00001 per cent per hr

or 1 per cent per 11 years), where weight is of secondary importance relative to long service life, as in stationary turbines

The time at which a stress can be sustained to failure is measured in a stress-rupture test and is normally reported as rupture values for 10, 100, 1,000, and 10,000 hr or more. Because of the higher stresses applied in stress-rupture tests, shown in Fig. 17.33, some extrapolation of data may be possible and some degree of uncertainty may ensue. Discontinuous changes at points *a* in the stress-rupture data shown in Fig. 17.34 are associated with a change from transgranular to intergranular fracture, and further microstructural changes can occur at increasing times. A composite picture of various high-temperature test results is given in Fig. 17.35 for a type 316 austenitic stainless steel.

(c) Material Selection

The aluminum and magnesium light alloys used in aircraft in their heat-treated condition have high-temperature applications limited to about 400°F. Low-density titanium alloys, of the alpha-beta heat-treatable type, have been used for aircraft gas-turbine compressor parts (where creep properties become important) in 600 to

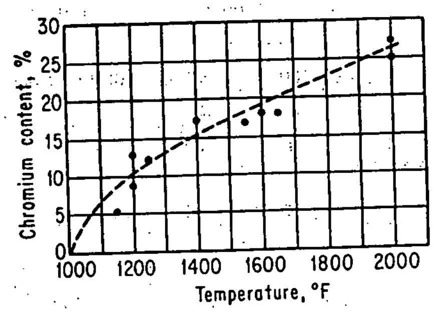


FIG. 17.36. The effect of chromium content on service temperature, without causing excessive sealing in alloys.[12]

1000°F applications. Low-alloyed steels are generally used safely to about 700°F in the high-temperature range. Above these temperatures, special alloys have been developed for the 1200 to 1800°F range and the 1800 to 2800°F range (the refractory alloys).

Alloys used in the temperature range of 1200 to 1800°F are essentially of the iron-, nickel-, and cobalt-base type and their structures, although predominantly of a single phase, may contain intermetallics and precipitating phases. Additions of chromium increase the oxidation resistance in a manner shown by Fig. 17.36. However, high-chromium alloys show depletion effects when used in a high vacuum. For nuclear applications, requiring low-neutron-absorption cross sections, alloys with high cobalt concentrations should be avoided. Some measured high-temperature stress-rupture properties for alloys used in this temperature range are listed in Table 17.3.

Applications of metals at the highest service temperatures call for the four refractory elements and their alloys. Each of these has a body-centered-cubic structure and a high melting point, oxidizes readily above 1200°F, and has the properties given in Table 17.4. The strength properties are affected by the impurities present (notably oxygen, carbon, and nitrogen), and therefore electron-beam melting under high vacuum is under investigation for attaining higher purities. Because of the oxidizing characteristics at high temperatures, protective coatings must be used.

Sheet products can be fabricated from Ta, Cb, and Mo by conventional means, since they have reasonably good ductilities at room temperature. However, tungsten is relatively brittle at room temperature and for this reason sheet-fabrication methods

Table 17.3. High-temperature Rupture Strength Properties for Some Superalloys[13]

Alloy	Principal alloy content, %	Rupture strengths, 1,000 psi				Typical applications
		1200°F		1500°F		
Fe base		100 hr	1,000 hr	100 hr	1,000 hr	
Incaloy 901	13 Cr; 43 Ni; 34 Fe, and Mo and Ti	94	78	24	15	Gas-turbine rotor disks
Refractory 26	18 Cr; 38 Ni; 20 Co, bal. Fe, and Mo and Ti	80	63	27	18	Gas-turbine parts, blading, bolting
		1200°F		1500°F		
Co base		100 hr	1,000 hr	100 hr	1,000 hr	
S816	20 Cr; 20 Ni; bal. Co and W and Mo and Cb and Fe	60	46	25	18	Jet-engine buckets
HS25	20 Cr; 10 Ni; 15 W, bal. Co	70	54	24	17	Jet-engine parts, sheet alloy
		1500°F		1800°F		
Ni base		100 hr	1,000 hr	100 hr	1,000 hr	
Inconel 713C	11.5 Cr; 74 Ni; Al and Mo and Cb	68	47	20	15	Jet-engine blades
Hastelloy R235	15.5 Cr; 10 Fe bal. Ni and Mo and Ti	40	30	8	5	Gas-turbine Jet-engine parts, sheet
Rene 41	19 Cr; 11 Co; 10 Mo; bal. Ni	45	29	11	...	Gas turbine, sheet, bolting

Table 17.4. Properties of Refractory Elements

Element	Melting point, °F	Density, lb cu in.	Recrystallization temp., °F	Young's modulus of elasticity at room temp., E, psi
Columbium, Cb.	4379	0.31	1785–2100	30,000,000
Molybdenum, Mo.	4730	0.369	2100–2200	47,000,000
Tantalum, Ta.	5425	0.60	2200–2400	27,000,000
Tungsten, W.	6170	0.697	2200–3000	50,000,000

require special development. Joining by fusion welding results in ductile joints for Ta and Cb, whereas for W and Mo grain-coarsening and cracking tendencies are factors of importance in attaining usable products.

High-temperature strength properties are substantially improved by alloying additions to the refractory elements. Tungsten-base alloys, some in the development stage, are used at the highest temperatures. Some typical rupture strength properties determined in heats are given in Table 17.5.

Table 17.5. Rupture Strength Properties[13]

Metal	Nominal alloy content, %	Rupture strength, 1,000 psi	
		2000°F	
Cb base FS82.	33 Ta, 7.5 Zr	10 hr	100 hr
		25	18
Mo base.	0.5 Ti	2000°F, 100 hr	2400°F, 100 hr
		34	10
		2500°F	
W base.	2% ThO_2	10 hr	100 hr
		29	22

17.13. MATERIALS FOR LOW-TEMPERATURE APPLICATION

Materials for low-temperature application are of increasing importance because of the technological advances in cryogenics. The most important mechanical properties are usually strength and stiffness, which generally increase as the temperature is decreased. The temperature dependence on ductility is a particularly important criterion in design, because some materials exhibit a transition from ductile to brittle behavior with decreasing temperature. Factors related to this transition are microstructure, stress concentrations present in notches, and the effects of rapidly applied strain rates in materials. Mechanical design can also influence the tendency for brittle failure at low temperature, and for this reason, it is essential that sharp notches (which can result from surface-finishing operations) be eliminated and that corners at changes of section be adequately filleted.

Low-temperature tests on metals are made by measuring the tensile and fatigue properties on unnotched and notched specimens and the notched impact strength. Metals exhibiting brittle characteristics at room temperature, by having low values of per cent elongation and per cent reduction in area in a tensile test as well as low impact strength, can be expected to be brittle at low temperatures also. Magnesium alloys, some high-strength aluminum alloys in the heat-treated condition, copper-beryllium heat-treated alloys, and tungsten and its alloys all exhibit this behavior. At best, applications of these at low temperatures can be made only provided that they adequately fulfill design requirements at room temperature.

When metals exhibit transitions in ductile-to-brittle behavior, low-temperature applications should be limited to the ductile region, or where experience based on field tests is reliable, a minimum value of impact strength should be specified. The failure,

by breaking in two, of 19 out of 250 welded transport ships in World War II, caused by the brittleness of ship plates at ambient temperatures, focused considerable attention on this property. It was further revealed in tests that these materials had Charpy V-notch impact strengths of about 11 ft-lb at this temperature. Design specifications for applications of these materials are now based on higher impact values. For temperatures extending from subatmospheric temperatures to liquid-nitrogen temperatures (−320°F), transitions are reported for ferritic and martensitic steels, cast steels, some titanium alloys, and some copper alloys.

Design for low-temperature applications of metals need not be particularly concerned with the Charpy V-notch impact values provided they can sustain some shear deformation and that tensile or torsion loads are slowly applied. Many parts are used successfully in polar regions, being based on material design considerations within the elastic limit. When severe service requirements are expected in use, relative to rapid rates of applied strain on notch-sensitive metals, particular attention is placed on selecting materials which have transition temperatures below that of the environment.

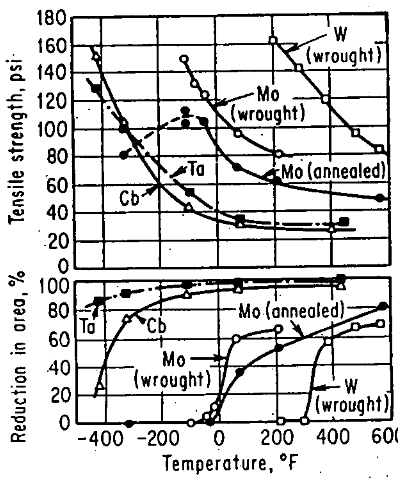


Fig. 17.37. Strength and ductility of refractory metals at low temperatures.[13]

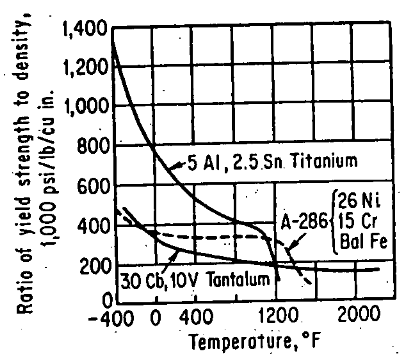


Fig. 17.38. Yield-strength-to-density ratios related to temperature for some alloys of interest in cryogenic applications.[13]

Some important factors related to the ductile-to-brittle transition in impact are the composition, microstructure, and changes occurring by heat-treatment, preferred directions of grain orientation, grain size, and surface condition.

The transition temperatures in steels are generally raised by increasing carbon content, by the presence of more than 0.05 per cent sulfur, and significantly by phosphorus at a rate of 13°F per 0.01 per cent P. Manganese up to 1.5 per cent decreases the transition temperature and high nickel additions are effective, so that in the austenitic stainless steels the behavior is ductile down to liquid-nitrogen temperatures. In high-strength medium-alloy steels it is desirable, from the standpoint of lowering the transition range, that the structure be composed of a uniformly tempered martensite, rather than containing mixed products of martensite and bainite or martensite and pearlite. This can be controlled by heat-treatment. The preferred orientation that can be induced in rolled and forged metals can affect notched impact properties; so that specimens made from the longitudinal or rolling direction have higher impact strengths than those taken from the transverse direction. Transgranular fracture is normally characteristic of low-temperature behavior of metals. The metallurgical factors leading to intergranular fracture, due to the segregation of

Table 17.6. Test Properties

Tensile Test Properties

Material	Tensile strength, 1,000 psi		Yield strength, 1,000 psi		Elongation, %		Modulus of elasticity, 100 psi	
	80°F	−100°F	80°F	−100°F	80°F	−100°F	80°F	−100°F
Beryllium copper 2% Be—heat-treated sheet...	189	194	154	170	3	3	19.1	19.1
Phosphor bronze 5% Sn; spring temp...........	98	107	90	97	7	11	16.5	17.1
Molybdenum sheet partly recrystallized.........	97	141	75	130	19	3.5	48.7	47.5
Tungsten wire, drawn....	214	...	196	...	2.6	....	56	
Tantalum sheet, annealed	55	71	36	66	27	25	28.2	28.3
Nilvar sheet 36 Ni; 74 Fe, rolled................	76	101	62	76	27	30	21.8	19.7

Sheet Fatigue Properties

Material	Fatigue strength at 2 × 10⁷ cycles 1,000 psi		Endurance ratios	
	80°F	−100°F	80°F	−100°F
Beryllium copper..............	31	45	0.16	0.23
Molybdenum....................	46	65	0.59	0.47
Tantalum......................	35	41	0.64	0.58
Nilvar........................	26	30	0.33	0.30

Charpy V-notch Impact Strength

Material	Impact strength, ft-lb	
	80°F	−100°F
Beryllium copper..............	5.4	5.4
Phosphor bronze...............	46	44
Nilvar........................	97	77

embrittling constituents at grain boundaries, cause concern in design for low-temperature applications. In addition to the control of these factors for enhanced low-temperature use, it is important to minimize or eliminate notch-producing effects and stress concentration, by specifying proper fabrication methods and providing adequate controls on these, as by surface inspection.

Examples of some low-temperature properties of the refractory metals, all of which have body-centered-cubic structures, are shown in Fig. 17.37. The high ductility of tantalum at very low temperatures is a distinctive feature in this class that makes it attractive for use as a cryogenic (as well as a high-temperature) material. Based on the increase of yield-strength-to-density ratio with decreasing temperatures shown in Fig. 17.38, the three alloys of a titanium-base Al (5) Sn (2.5), an austenitic iron-base Ni (26), Cr (15) alloy A286, as well as the tantalum-base Cb (30), 10 V alloy, are also useful for cryogenic use.

Comparisons of the magnitude of property changes obtained by testing at room temperature and −100°F for some materials of commercial interest are shown in Table 17.6.

17.14. RADIATION DAMAGE[31]

A close relationship exists between the structure and the properties of materials Modification and control of these properties are available through the use of various metallurgical processes, among them the concept of nuclear radiation. Nuclear radiation is a process whereby an atomic nucleus undergoes a change in its properties brought about by interatomic collisions.

The energy transfer which occurs when neutrons enter a metal may be estimated by simple mechanics, the quantity of energy transferred being dependent upon the atomic mass. The initial atomic collision, or primary "knock-on" as it is called, has enough energy to displace approximately 1,000 further atoms, or so-called secondary knock-ons. Each primary or secondary knock-on must leave behind it a resulting vacancy in the lattice. The primaries make very frequent collisions because of their slower movement, and the faster neutrons produce clusters of "damage," in the order of 100 to 1,000 angstroms in size, which are well separated from one another.

Several uncertainties exist about these clusters of damage, and because of this it is more logical to speak of radiation damage than of point defects, although much of the damage in metals consists of point defects. Aside from displacement collisions, replacement collisions are also possible in which moving atoms replace lattice atoms. The latter type of collision consumes less energy than the former.

Another effect, important to the life of the material, is that of transmutation, or the conversion of one element into another. Due to the behavior of complex alloys, the cumulative effect of transmutation over long periods of time will quite often be of importance. U^{235}, the outstanding example of this phenomenon, has enough energy after the capture of a slow neutron to displace one or more atoms.

Moving charged particles may also donate energy to the valence electrons. In metals this energy degenerates into heat, while in nonconductors the electrons remain in excited states, and will sometimes produce changes in properties.

Figure 17.39 illustrates the effect of irradiation on the stress-strain curve of iron crystals at various temperatures.[32] In metals other than iron irradiation tends to produce a ferrous-type yield point and has the effect of hardening a metal. This hardening may be classified as a friction force and a locking force on the dislocations. Some factors of irradiation hardening are:

1. It differs from the usual alloy hardening in that it is less marked in cold-worked than annealed metals.
2. Annealing at intermediate temperatures may increase the hardening.
3. Alloys may exhibit additional effects, due for example to accelerated phase changes and aging.

The most noticeable effect of irradiation is the rise in transition temperatures of metals which are susceptible to cold brittleness. Yet another consequence of irradiation is the development of internal cracks produced by growth stresses. At high enough temperatures gas atoms can be diffused and may set up large pressures within the cracks.

Some other effects of irradiation are swelling, phase changes which may result in greater stability, radiation growth, and creep. The reader is referred to ref. 31 for an analysis of these phenomena.

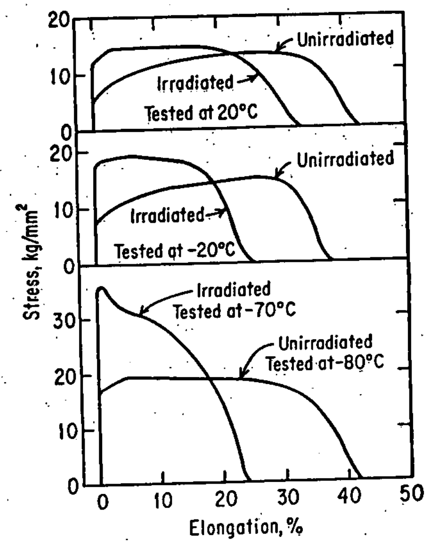


FIG. 17.39. Effect of irradiation on stress-strain curves of Fe single crystals tested at different temperatures. Irradiation dose 8×10^{17} thermal n/cm². (*Courtesy of ref. 31.*)

17.15. PRACTICAL REFERENCE DATA

Table 17.7 through 17.10 give various properties of commonly used materials. Figure 17.40 provides a hardness conversion graph for steel. References 21 through 30 yield more information.

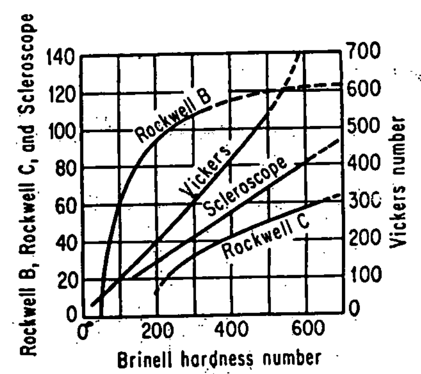


FIG. 17.40. Hardness conversion curves for steel.